GAS TURBINE MATERIALS

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SIMPLE OIL-ENGINED PLANT OPERATION
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"THE OIL ENGINE AND GAS TURBINE" WORLD DIRECTORY OF MANUFACTURERS
"THE OIL ENGINE" MANUAL

GAS TURBINE MATERIALS

A SURVEY OF HIGH-TEMPERATURE MATERIALS
AND THEIR APPLICATION TO THE GAS TURBINE

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AND
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FOREWORD

THE GAS TURBINE is essentially a high-temperature engine and the development of special materials to meet the requirements of high efficiency and reliable operation in association with high temperatures is without doubt one of the outstanding metallurgical achievements of all time. While much has been done by the engineer to ease the problem by such expedients as swept cooled surfaces and by improved design arising from a better understanding of the behaviour of components in service, there remains nevertheless an insistent demand for new materials of better strength/weight ratio, of high resistance to corrosion at elevated temperatures, of high residual ductility after operating for long periods at relatively low stresses, of high thermal conductivity and low thermal expansion and in fact, an embarrassing combination of these and other requirements.

In common with all new developments the gas turbine has suffered considerable change in fortune since its original development for aircraft propulsion, and many of the successes originally predicted for its application to industrial requirements have not obtained in practice. Without doubt its first and finest use still relates to aircraft propulsion but there is a healthy, though limited, application for industrial purposes. Particularly is this true where oilfield gas is available as fuel, since this enables a relatively cheap and reliable machine to be used with advantage. Although a great amount of development work has been done regarding its application to vehicle propulsion, it has not so far found much favour in this respect.

It has become clear in recent years that while there may still be a market for small industrial gas turbines of up to 5000-kW output, there is a growing demand in Canada and elsewhere for much larger units. Accordingly, if companies are prepared to specialise in the design and manufacture of gas turbine machinery, the outlook is healthy and no company engaged in the design and production of turbines can afford to overlook this aspect of the power business.

The excellent survey which has been made by the Authors of this book on the manufacture, properties and use of special materials for gas turbines, will be read with considerable interest and will, without doubt, be found most helpful in the use and selection of special materials by designers and users alike.

A. T. BOWDEN, Ph.D., B.Sc.(Eng.), M.I.Mech.E.

PREFACE

THIS BOOK is intended to be a survey and not a detailed treatment of the field of gas turbine materials. It is hoped that it will fill to some extent the gap between the metallurgical textbooks and those works concerned with the technology of design and performance of gas turbines.

Both of these subjects are specialist by nature and it is thought that the treatment of high-temperature materials from the aspect of their use in the gas turbine will render the book of particular value to the engineer, metallurgist and student seeking a unified review of gas turbine materials.

The growing importance in recent years of that aspect of metallurgy concerned with the provision of alloys for use in the gas turbine is considered to be a justification for including a summary of the development of such alloys. It may not always be fully appreciated how the thermodynamic design of the gas turbine affects the choice of materials; this is dealt with generally together with the influence of fuels on operating conditions and thus on material selection.

The working life of materials is set in a stress-temperature environment in which chemical influences may be prominent. This is considered in Chapter Three with emphasis on the phenomenon of creep which is intimately associated with stress-temperature effects. Materials for the individual components in the gas turbine are dealt with in more detail in Chapter Four from the aspect of the operating conditions they have to face.

Special materials of a non-metallic nature and metal/non-metal combinations may become increasingly important as the gas turbine develops, and these are considered in Chapter Five together with special treatments designed to extend the application of otherwise limited materials. Finally, manufacturing processes are reviewed in relation to the various components and the more important processes dealt with in detail.

It is not possible to give references to all the works which have been consulted but the debt of the authors to a wide range of published papers, articles and books is acknowledged. Where particular reference is made to a published work this is given at the end of the chapter, with, in some cases, a few additional selected references.

G. LUCAS, B.Sc., Ph.D., A.I.M.
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July 1957

ONE

An Outline of the Development of High-temperature Materials

THE ENGINEER today makes considerable use of materials and alloys capable of retaining strength at high temperatures, often under the influence of a corrosive environment. The development of materials specifically for their high-temperature properties, however, is of comparatively recent origin. It has been closely linked with the requirements of industry; the steam plant together with the internal-combustion engine and chemical process plant have all exerted considerable influence, while during the last 15 years some outstanding metallurgical advances in the heat-resisting alloys have been stimulated by the advent of the gas turbine as an important source of power.

Early industrial processes did not in the main require the use of high temperatures and pressures. Consequently there was little metallurgical interest in the effect of temperature on metals until comparatively late in the nineteenth century. When the initial development period of plant and processes had been sufficiently long for operating experience to accumulate and for engineers to acquire mastery of design techniques, the permissible pressures and temperatures rose fairly rapidly. After the invention of the steam engine working pressures increased slowly above the original level of about 10 lb/in^2 and for about a century the temperatures associated with steam plant were not sufficiently great to cause concern over the strength of the materials of construction. Indeed the metals probably exhibited greater tensile strength at the moderate temperatures then employed, than they had at room temperature. About 1885 the first signs of the end of this phase were to be seen in the increased attention given by metallurgists to the effect of temperature on the physical properties of metals. Since that time the availability of suitable materials with the properties of heat and corrosion resistance has been one of the most significant factors controlling engineering progress. The introduction of the technique of superheating steam in boilers, which began to be adopted at the close of the nineteenth century, caused operating temperatures to rise more rapidly. By 1920 operating pressures of 350 lb/in^2 and temperatures up to 400°C (750°F) were in general use; these temperatures were close to the limit of the available

alloys and it is noteworthy that in the next 10 years, while pressures of 600 lb/in^2 had been generally attained with instances of 900 lb/in^2 and above, temperatures had increased only to about 450°C (840°F). Even at this time there was insufficient data concerning the properties of materials under these conditions and the best of the available alloys did not allow sufficient margin for safety to warrant the use of much higher temperatures.

An important use for high-temperature materials lay in the oil industry. On the whole pressures in refinery plant were lower than in the steam plant but temperatures were often considerably higher. In the 1920s upwards of 20 separate oil cracking processes were available, most of which operated at about 100 lb/in^2 and in the 400–450°C (750–840°F) region. In a few cases pressures up to 600 lb/in^2 and temperatures of 480°C (900°F) were employed while one process operated at 540°C (1000°F). The desire to obtain a high yield of petrol from the petroleum was one reason for the use of high temperatures and constituted the origins of a further demand for superior materials, for the safety factor was probably lower here than in other fields. Lack of precise long-time strength data, however, made it difficult to assess the degree of closeness to failure in service which existed in such applications.

High temperatures were also encountered in the internal-combustion engine and the aircraft exhaust gas turbo-supercharger. Exhaust valves, for instance, were exposed to very arduous conditions, for the gas leaving the cylinder was of the order of 600°C (1110°F). The highest pressures were in use in chemical plants. Production of synthetic ammonia by the Haber process required up to 200 atmospheres with, in 1926, temperatures of 600°C (1110°F), while the Claude process employed 700 atmospheres and 550°C (1020°F).

The use of high pressures presented a much easier problem than the use of high temperatures, for in many instances stress due to pressure could be kept at suitably low levels by the designer. Provision of sufficient metal area to carry the load at low stress intensity was in many cases attended by no great difficulty and thus the reduction in material strength which occurred rapidly above about 400°C (750°F) could be rendered less limiting in its effect. It is this fact which made it possible to employ much the same materials in cases where the operating temperature might differ by over 100°C but with much the same factor of safety. This was particularly the case in the chemical plants where the combination of the effects of stress and temperature were less acutely felt in comparison, than in the prime movers where the effects of centrifugal force increased as the mass of the rotating parts. In the latter it

was always the combination of stress and temperature which was the limiting factor in the selection of operating temperatures.

In these diverse uses for materials with good high-temperature characteristics the different operating conditions called for various combinations of properties. The exhaust valve needed good impact resistance and the ability to maintain its shape under fluctuating temperature as well as good fatigue and scaling resistance. In chemical process plant the resistance to attack by acids or gases such as hydrogen and nitrogen was required. Steam turbine blades had to be able to resist the erosion produced by the impact of drops of water suspended in the steam. Thus before the gas turbine created its own particular type of operating conditions classes of materials had developed in which the particular properties required were reflected in the alloys. Strength at high temperature and corrosion resistance were however common to them all.

Early Iron-base Heat-resisting Materials

Although there is an extremely wide range of alloys based on iron, in which use is made of particular elements to confer specific properties, two elements are outstanding in their influence on the heat and corrosion resistance of these alloys. They are the elements chromium and nickel. In the pure state both possess a melting point similar to iron (chromium 1830°C (3326°F), nickel 1452°C (2646°F), iron 1535°C (2795°F)). Chromium was first isolated in the metallic state by Vauquelin (France, 1798) and later considered as a suitable element to alloy with iron by Faraday. The development of the first Fe-Cr alloy is however ascribed to Berthieu (France, 1821) which resulted in their hardness and acid-resisting properties being recognised, while Faraday and Stodart (England, 1822) noted their oxidation resistance. Some years elapsed however before the importance of the chromium irons and steels began to be fully realised. Baur (U.S.A., 1865) was concerned with the production of chromium irons and steels in his experiments and patents but the first systematic development of Fe-Cr alloys probably took place in France. Brustlein (1876) is linked with this development while early research into the influence of carbon additions on Fe-Cr alloys was carried out by Carnot and Goutal (France, 1898).

The constitution of both Fe-Cr and Fe-Cr-C alloys is such that a wide range of characteristics can be developed, resulting in varied physical properties in the material, depending on the amounts of chromium and carbon present. Investigations into the properties of these alloys were carried out by Hadfield (Great Britain, 1892), who observed that 9% chromium steels containing carbon, silicon and manganese had remarkably good acid resistance, and Monnartz

(France, 1911) who mentioned that Fe-Cr steels with low carbon content also possessed the property of acid resistance. High-chromium steels also received attention. Strauss and Maurer (Germany, 1909/10) considered steels with 20% chromium, but the outstanding advance in the development of these alloys was due to Brearley (Great Britain, 1912–14). In the course of experiments on steels capable of resisting gaseous erosion in gun-barrel linings he noted the features of high melting point and resistance to corrosion possessed by the iron-chromium steels. This resulted in his discovery of the "stainless steels" in which the properties of the alloys were developed by a hardening and tempering process. His subsequent patent specified 9–16% chromium with less than 0·7% carbon, and the use of heat treatment to suppress free carbide formation. Other alloying elements were tried in conjunction with chromium, a notable one being silicon. Armstrong (U.S.A., 1912/20) became aware of the influence of silicon in improving the iron-chromium alloys and from his work the well-known range of "Silchrome Steels" stemmed, in which the silicon gave improved oxidation resistance at high temperatures.

Although Cronstedt isolated nickel in 1751 and Faraday and Stodart in 1820 reported its use as an alloying element with iron, the importance of nickel as an alloying element in steel arose after chromium steels had been developed to the stage of commercial use. Marbeau (France, 1885) was connected with the production of a range of iron-nickel alloys and nickel steels and these attracted the attention of Riley (Great Britain, 1889) who investigated these alloys and published his paper on "Alloys of Nickel and Steel" (1). This showed that by the addition of nickel greater tensile strength and a higher elastic limit could be obtained without sacrificing ductility and indicated the non-corrodibility of the high-nickel steels. The commercial value of these alloys soon became apparent since nickel had recently become available in greater quantity, as well as more cheaply, and the nickel steels rapidly found increasing use.

Steels using both nickel and chromium were investigated in France (1902) and their oxidation resistance led to the iron-nickel-chromium alloys being considered for pyrometry applications by Strauss (Germany, 1912) and Monnartz (France, 1911) mentioned nickel-chromium alloys for acid resistance. The austenite-forming tendency of nickel was recognised and two main types of nickel-chromium steels distinguished; the ferritic type containing up to about 2% nickel and the austenitic with considerably higher nickel content. Krupps (Germany, 1912) patented two steels characteristic of each of these. The ferritic steel (0·15% C-14% Cr-1·8% Ni)

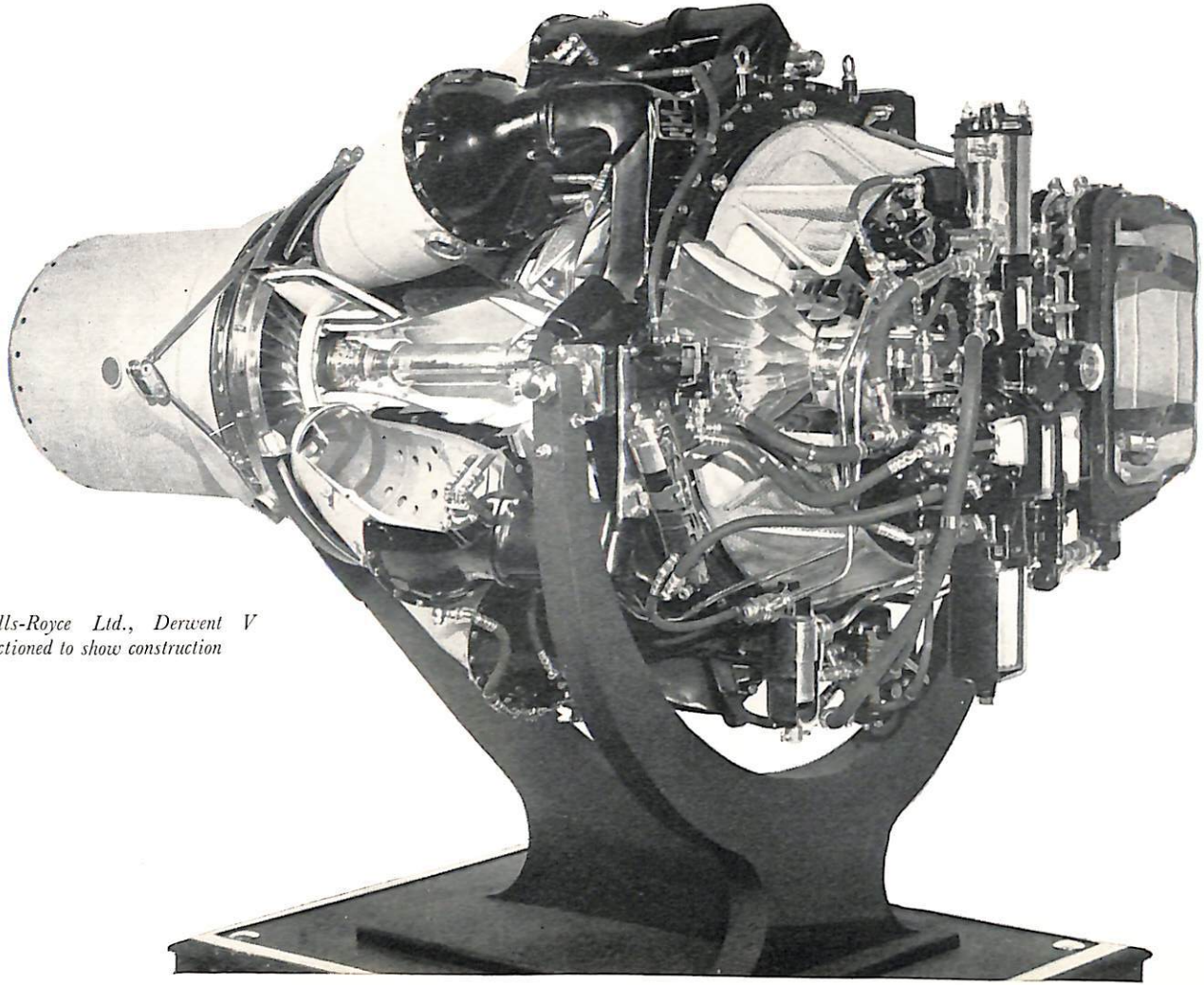


Plate 1. *Rolls-Royce Ltd., Derwent V turbo-jet, sectioned to show construction*

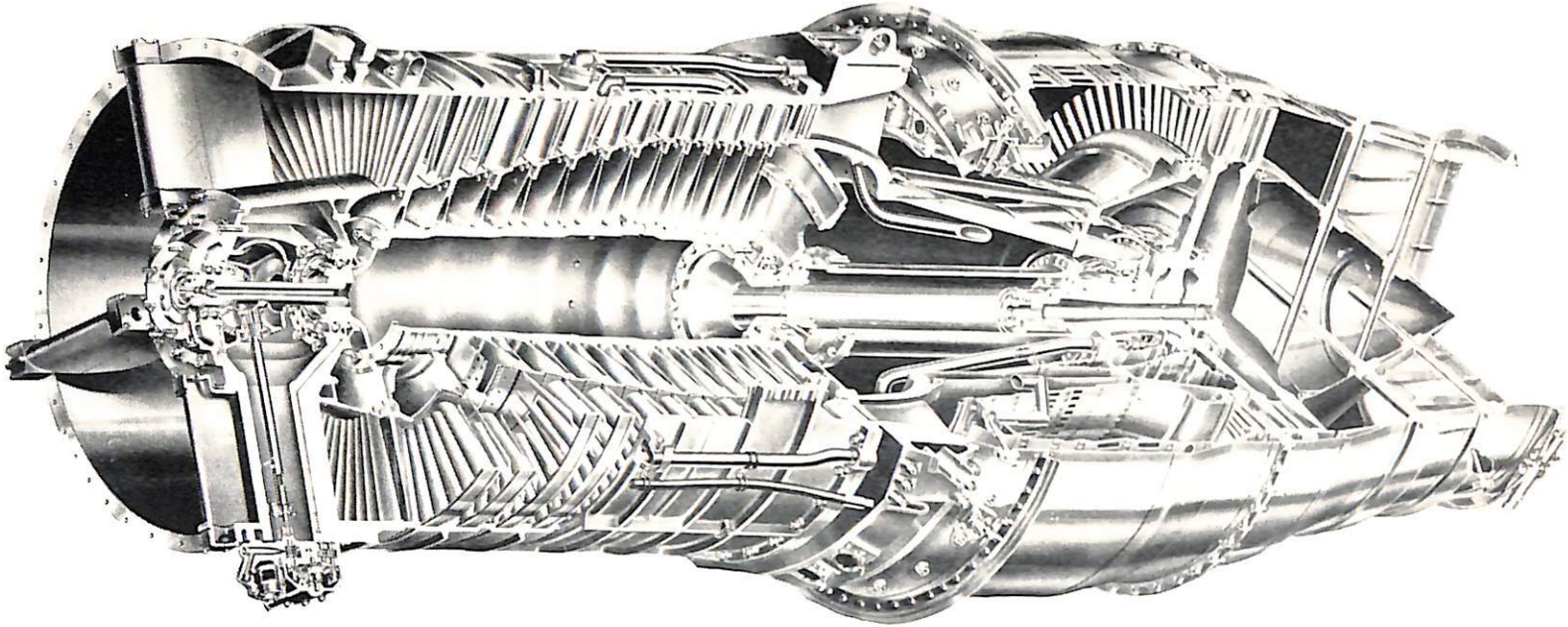


Plate 2. Perspective sectional view of the Armstrong Siddeley Sapphire A.S.Sa.6 turbo-jet

was representative of its class while the austenitic (0·25% C-20% Cr-7% Ni), was the forerunner of the now well known 18% Cr-8% Ni alloy possessing good high temperature strength and oxidation resistance. A range of these high-chromium–nickel and high-chromium–high-nickel steels was produced, notably in France about 1917, and found a variety of high-temperature applications, such as steam turbine blading and rotors, exhaust valves and parts of synthetic ammonia plants. In all these applications good high-temperature strength and stability were needed as well as corrosion or oxidation resistance.

Early Cobalt-base Alloys

Originally considered for tool steel applications these alloys were first developed by Haynes in the U.S.A. about 1910 onwards. Although their hot hardness had been one of the characteristics which made them suitable for metal cutting, the associated creep strength was not at first recognised. It was thus not until the 1930s that their importance as high-temperature materials came to be realised.

The early cobalt-base alloys were alloyed with chromium, with tungsten, molybdenum and manganese as minor additions but tungsten came to be used in quite large amounts when it became known that amounts above 3% greatly increased the hardness of the alloy. Since tungsten forms a continuous series of solid solutions with the cobalt-chromium alloy, very hard materials could be obtained by increasing the tungsten content. An increase in the elastic limit could be conferred by the use of tungsten in the presence of small amounts of boron, carbon or silicon and the brittleness, usually associated with cast materials, could be reduced.

Typical alloys contained about 15% Cr with 10–40% W, in which the hardness steadily increased with the tungsten content. With the increased tungsten content, however, the forgeability is greatly reduced and the alloys are preferably cast. Molybdenum has much the same effect as tungsten, although larger amounts of molybdenum are needed to produce an equivalent increase in hardness.

Temperature and the Physical Properties of Metals

There was little metallurgical interest in the effect of temperature on materials before 1885 but with the trend towards the use of elevated temperatures in industrial plant the subject began to acquire some importance. Subsequently investigators such as Rosenhain and Humphrey, Le Chatelier, Martens, Charpy, Brinell and Unwin, to mention only a few of the better known names, gave their attention to establishing for current materials over a range of temperatures, the same physical properties that could normally be

measured by room-temperature tests. Brinell (England, 1905) carried out hardness tests at high temperatures, and Le Chatelier (France) was responsible for some early high-temperature tensile tests on various alloys.

During this period of high-temperature testing results often differed widely, primarily because there was no established testing technique and each investigator solved the problems of temperature and loading control in the way which seemed best to suit the needs of the case. The available extensometers had been developed for room-temperature testing and in any case were not capable of a high degree of accuracy. Between 1910 and 1925 difficulties had been to a large extent overcome and improved techniques and better extensometers had allowed the accumulation of reliable data on the variation of the physical properties of metals with temperature. Tensile and impact tests, alternating stress tests, torsion and hardness tests and x-ray spectroscopy were used to provide this data. It became known that both ultimate strength and elastic limit decreased with temperature while total elongation and reduction of area increased. There were qualifications to these generalisations for in some instances there was a temperature range between about 250°C and 425°C (480–795°F) in which the trends were reversed, with tensile strength being greater and the reduction of area and elongation less than the room-temperature values. This phenomenon was known as "Blue Brittleness".

Some experiments carried out by Bach (2) in 1903 are of interest in that they foreshadow some important conclusions which were drawn later. Specimens were tested at room temperature, 200, 300, 400, 500 and 550°C. The results showed the trends outlined above; the tensile strength increased up to 300°C (572°F) then decreased, while the total extension decreased up to 200°C (392°F) and then increased. Bach observed, however, that while increased duration of action of the load had no effect at ordinary temperatures it caused a progressive decrease in strength at 300, 400 and 500°C (572, 752 and 932°F) respectively. Furthermore he stated that at 500°C (932°F) the long-time figures for extension and reduction of area were considerably lower than the corresponding short-time test values. These observations anticipated by a good many years the important conclusions of Dickenson (3) (1924) that only long-time creep testing was of any value in deciding the working stresses a material could safely bear over a long period.

The Phenomenon of Creep

Although the phenomenon of creep in metals had been known in the mid-nineteenth century the first important research was that of

Andrade (1910–14). This demonstrated that plastic flow occurred in pure metals and alloys under a steady continuous load. Chevenard (France, 1919) extended the range of materials investigated to nickel and the nickel alloys and published (4) the first short-time creep test results; these indicated the relation of creep strain to temperature and stress and showed the superior resistance to creep of the Cr-Ni and Cr-Ni-W alloys in the 550–650°C (1020–1200°F) range. The implications of these researches were not at first fully appreciated for short-time tests continued to be made for some years and design stresses calculated on the basis of short-time data, albeit with an adequate safety factor. In stress calculations the assumption is usually made that the material is perfectly elastic, provided the stress does not exceed the elastic limit. If plastic flow takes place, stress relief occurs at the high-stress regions so that the actual stress becomes less than the elastic stress. Strain hardening accompanies the plastic flow and ultimately a brittle fracture may take place. The researches of Dickenson (1924) and Lea (5) (1924) drew attention to the importance of creep from the engineering aspect and showed that the ordinary short-time tensile test could not be relied on to indicate the stress a material could safely carry for long periods at high temperatures, for failure would take place under a smaller steady load than that required to fracture a specimen in the short-time tensile test.

Early Creep-resisting Materials

The early investigations, stimulated by the recognition of the importance of creep strength in deciding the suitability of materials for use at high temperatures, were concerned with the existing alloys. It was not until the available materials had been assessed for creep strength that attempts were made to improve materials. French and Tucker (6) (1924) distinguished the following types of alloys with superior creep resistance:

1. Stainless irons.
2. Austenitic stainless steels.
3. 11–14% Cr steels.
4. 8/25% Cr-35/80% Ni-Fe alloys.
5. Cr-Ni-Fe alloys with W or Mo additions.
6. Nickel base alloys with Cr or Al.
7. Cobalt base alloys with high Cr.

Several of these alloys were well known and in general use but some were relatively new and not considered to be of commercial application owing to their high cost. There was at the time a widely held

opinion that alloys which were not composed largely of iron were uneconomic, especially if considerable quantities were required. Some figures of comparative costs are given in Table I. This shows that although the highly alloyed materials were comparatively expensive in 1929 the best creep-resisting materials were much more so in 1950.

TABLE I

Comparative costs of high-temperature materials

MATERIAL	1929	1950
Low-carbon steels	1	1
Low-alloy steels	2	2–3
Stainless and heat-resisting steels	10–15	8–15
Nimonic 80	—	28
Nimonic 80A	—	33
Nimonic 90	—	50

The position with regard to the provision of high-strength, heat-resisting materials in 1924 was well summarised by French and Tucker. They said "the composition and treatment of steels may be varied to meet specific requirements at temperatures above 260°C (500°F). However, it is also evident from the described data that all steels 'weaken' with considerable rise in temperature. While this weakening may be delayed by additions of relatively large proportions of such elements as chromium . . . steels would not generally be expected to stand up under fairly high stress at temperatures exceeding about 650°C (1200°F). Thus for services at higher temperatures it would appear necessary to seek alloys in which the iron plays a secondary role instead of forming the largest part of the product."

In the same paper a summarisation of experience with the direct synthetic ammonia process was quoted. This amplified the statement of the trend towards the use of the highly alloyed materials and taken together the two provide an accurate forecast of the type of high-temperature material which is at present in use. J. S. Vanick said in this summarisation: "Pressures which can be held at ordinary temperatures exert stresses at the operating, elevated temperatures that would correspond to or exceed the limit of elasticity for practically all steels and ferrous alloys. So important has high-temperature strength become for this service that the property of resistance to corrosion which has thus far been associated with it

may be subordinated; partly because most of the elements which possess the property of strength or coherence at elevated temperatures also possess, in this case, a superior resistance to corrosion or deterioration. New investigations are leading into alloys of the inelastic type; alloys which possess a yield point at ordinary temperatures that closely approaches the ultimate strength. At elevated temperatures some plasticity would be expected which would not necessarily imply elasticity; at present these alloys are expensive, unmachinable, must be cast to shape with the difficulty that attends viscous fusions, and in the present state of development improve the high-temperature strength very slightly. Work on these important alloys will clear up many of the obstructions now encountered in their preparation and lead to new developments in the strength-at-high-temperature field."

The adoption of the casting alloys for high-temperature use was thus foreshadowed as well as the development of the highly alloyed type of non-ferrous materials such as the Nimonic alloys. A tremendous amount of research was, however, needed, for in the case of the newer materials quoted by French and Tucker there was still a lot to be learned about the influence of impurities and small changes in composition; casting and forging techniques were also capable of considerable development in the case of the higher melting point alloys. Inadequate knowledge of the factors affecting creep strength was evidenced by the dissimilar results of creep tests on apparently identical specimens. Although this was attributed in some measure to the variations in testing technique it was also realised that it could be due to alterations in the nature of the material arising from slight composition differences or a non-uniform heat or mechanical treatment prior to testing.

In the 10 years following the recognition of the need for creep testing of high-temperature materials many creep investigations were made and the field of available alloys widely surveyed. Some progress was made in extending the knowledge of the factors influencing creep strength and the range of usefulness of available alloys became well known. By 1930 it was accepted that existing ferritic steels were not suitable for use at temperatures in excess of 450°C (840°F). Although heat treatment and hot or cold mechanical work could be employed to improve the structure of the metal, the greater strength which this conferred could not be retained at temperatures above 450°C (840°F). At this level the influence of composition becomes very important and at working temperatures greater than 500°C (930°F) high strength is obtained primarily by the correct choice of composition.

The high-nickel–high-chromium austenitic steels, therefore,

gradually became established as the materials which possessed the most outstanding creep resistance and structural stability. These, however, suffered from the disadvantage that they exhibited a tendency to grain boundary embrittlement. This led to the use of tungsten and molybdenum additions as a means of reducing this tendency and later the nickel-molybdenum alloys of the Hastelloy type were developed for their resistance to this type of fault. Silicon and vanadium were also found useful as additions, the former for its beneficial effect on oxidation resistance while the latter promoted fine grain size and influenced carbide precipitation.

Although the austenitic steels were now known to exhibit superior creep resistance several factors combined to focus interest on the development of ferritic steels. The latter alloys were appreciably cheaper to manufacture and much easier to work, since they did not have the tendency to work harden to the same extent as the austenitic alloys. Furthermore the susceptibility to embrittlement was not present to the same extent. For the manufacture of large forgings, such as steam turbine rotor discs, a ferritic steel thus possessed distinct advantages and the same considerations later led to the use of ferritic discs in the aircraft gas turbine wherever possible. As the influence of molybdenum, manganese, vanadium, chromium and nickel became better known, a range of ferritic steels was developed in which the high-temperature properties were largely conferred by the use of these alloying elements. Molybdenum and vanadium were used to give high creep strength while oxidation resistance and hot hardness were conferred by suitable additions of chromium, nickel and manganese. In the U.S.A. the silicon-chromium steels had become well known for their oxidation resistance and moderate high-temperature strength, although their creep resistance was not outstanding. The main types of ferritic heat-resisting steels available in the 1930s, therefore, were:

1. Fe-C-Mo,
2. Fe-C-Mo-Cr,
3. Fe-C-Mo + Mn, Cr or Ni,
4. Fe-C-V + Mn, Cr or Ni,
5. Fe-C-Cr + W or Mo,
6. Fe-Si-Cr,

and it was from these that the ferritic gas turbine steels later came to be selected.

The austenitic steels were used in applications in which hot strength and resistance to oxidation at the higher temperatures was necessary, such as in steam turbine blading and internal-combustion engine exhaust valves. In the latter case a good hot

impact strength, combined with thermal shock resistance was essential. A wide range of alloys was tried over the years for exhaust valves, including very complex high-nickel chromium alloys with silicon and manganese and in some instances a small percentage of cobalt; molybdenum, vanadium and tungsten additions were also used, separately or together. The Stellites which had been introduced in the U.S.A. were employed to face the valves since their hardness was little affected by temperatures up to 1000°C (1830°F). A typical Stellite composition for this application was

% C	% Co	% Cr	% W
2·5/2·7	40/50	25/30	15/20

Although creep strength was seldom considered in the exhaust valve materials, a good deal of the research work undertaken in the development of these alloys was later applicable in the production of gas turbine alloys. An important instance is the knowledge of the manner in which carbide precipitation in these materials could be controlled. Many of the early attempts to develop creep resisting gas turbine alloys were based on available internal-combustion engine exhaust valve and steam turbine blading materials in which the influence of the carbide formers, such as tungsten, columbium, vanadium and titanium, in the presence of chromium and nickel had become known.

It was during this period that the first investigations were begun to find materials with creep properties which would enable them to withstand higher operating temperatures than were currently in use. A notable instance was the research carried out at the National Physical Laboratory (7) on the nickel-chromium and the iron-nickel-chromium alloys. Here the purpose was to study the systems likely to yield materials with great strength in the 650–800°C (1200–1470°F) range and suitable for future requirements. The results indicated that nickel chromium alloys with suitable additions had excellent possibilities, a composition

% Cr	% Ni	% Fe	% W	% C	% Si
30	30	33·5	4	1·5	1·0

in particular having a good rupture strength at 800°C (1470°F) being superior to the industrial steels then in use. The implication that high-cost materials of this type, with less than one-third iron, were representative of the future heat-resisting alloys was widely deplored at the time although it had been recognised by some that the solution to the problem lay in this direction.

The work on the nickel-base alloys is of interest in view of the later importance of the Nimonic alloys. This showed a 70%

Ni-30% Cr alloy to have the best properties at 800°C (1470°F) although it was not as good at 650°C (1200°F) as the other low-chromium alloys. This result foreshadowed the experience in the development of Nimonic 75, where the same phenomenon was encountered. It was observed that all the alloys investigated which had good high-temperature properties lacked ductility and were not easy to work even at high temperatures. It was also observed that the cast form, where the metal could be satisfactorily cast, had better creep properties than the forged material, although the latter had a high-temperature ultimate strength which was closer to the room-temperature value. In some instances in the N.P.L. investigations it had been necessary to vacuum cast, a procedure which would never have been adopted for commercial production at that period but which was ultimately realised to be inevitable with some materials if consistent physical properties were to be attainable. As early as 1930 therefore the future problems and trends had been outlined, confirming the previous observations of French and Tucker.

THE INFLUENCE OF THE GAS TURBINE ON MATERIAL DEVELOPMENT

Great Britain

The first Whittle gas turbines operated at a temperature which was based on the limitations imposed by the high-temperature strength of the existing materials, bearing in mind the need to keep this temperature as high as possible in order to maintain a reasonable efficiency. Available materials suitable for the blades and discs were the Firth-Vickers alloys, Rex 78 and Stayblade. The latter, an 18% Cr-8% Ni type with titanium and aluminium additions had been available for some years as a steam turbine blade material known as Staybrite, while Rex 78, produced specifically for the gas turbine, was a complex alloy of the Ni-Cr-Fe type with a number of additions. The creep strength of Rex 78 was markedly better than Stayblade and was subsequently further improved by the development of a more suitable heat treatment, but 650°C (1200°F) was about the limit at which it could be used. The small reserve margin of high-temperature strength contributed to a number of blade and disc failures experienced with the early gas turbines and this, combined with recognition of the fact that substantially higher operating temperatures were essential to further development of the gas turbine, resulted in a search for new materials.

The Nimonic Alloys

In 1939 the Air Ministry requested the Mond Nickel Co. to consider the possibility of producing an alloy which could operate at red heat under high stresses. At this time the chief need was for a material suitable for turbine rotor blades in existing gas turbine engines with a turbine inlet temperature of about 600°C (1112°F) but the figure of 750°C (1382°F) was quoted as the operating temperature in the initial specification for the new material.

The problem was tackled initially by a review of the existing data on high-temperature materials. As a result it was decided that an alloy based on the already well-known 80% Ni-20% Cr series of alloys offered the greatest promise. This type of alloy had been investigated at the National Physical Laboratory; some information was thus already available regarding its high-temperature properties and, furthermore, previous work had demonstrated the influence of tungsten, molybdenum, titanium, columbium and tantalum additions. From the production aspect a forgeable material was considered essential and adequate fabrication experience existed for this type of alloy. Oxidation resistance at high temperature was another requirement of the specification; the 80% Ni-20% Cr type possessed this property and had furthermore already been used for other high-temperature applications.

Early in 1940 a new alloy had been developed with the properties at 750°C (1382°F) called for in the specification. It became known as Nimonic 75. The general level of operating temperatures in the gas turbine had not risen as high as 750°C (1382°F), however, and was closer to 650°C (1200°F). At this lower temperature it was found that the alloy did not exhibit the same outstanding qualities that it possessed at the higher temperature and for this reason was not used as a rotor blade material, as had originally been intended. Temperatures in the turbine nozzles and combustion chamber flame tubes, however, were higher although stresses were lower and Nimonic 75 found an immediate use in these components. In the early years of the gas turbine development it had no serious competitor for this application.

The high-temperature properties of the alloy depended on the solution and subsequent precipitation of titanium carbide. Both aluminium and titanium were used as hardeners and the metal heat treated to produce its optimum high-temperature properties. Considerable experiment was needed to establish the correct amounts of these additions. The presence of aluminium renders forging difficult and the problem was to combine aluminium and titanium in the proportion which gave the required high-temperature strength and also allowed the material to be forged with reasonable ease.

Further research into the basic composition of the alloy continued in an attempt to produce a material with the hot strength needed in a rotor blade material. The addition of cobalt was tried and in 1942 Nimonic 80 was produced with a creep strength superior to all existing alloys. Early samples were produced by laboratory methods which, though suitable for the production of small quantities, were not satisfactory for large-scale supply. When large-scale production was begun considerable difficulties were experienced in maintaining the creep resistance of the alloy consistent as the presence of impurities had an extremely deleterious effect and precise control of composition was not easy to achieve. By 1942, however, Nimonic 80 had become a standard production material with guaranteed creep resistance. This was achieved by a system of batch creep testing since each furnace heat was found to have slightly different properties.

A variant produced from this alloy was Nimonic 80A, which had superior creep strength up to 750°C (1382°F) but could be used for short-time applications at temperatures up to 820–850°C (1508–1562°F). Furthermore its oxidation resistance was as good as that of Nimonic 80. Both of these alloys were adversely affected by cold work, which had the effect of lowering creep resistance and causing rupture to take place sooner and at lower loads.

Two advances were made in this range by the use of cobalt to replace some of the nickel content. Nimonic 90 and 95 became available in 1950 and 1952 and in both of these alloys 16% of cobalt was used to improve the high-temperature creep strength. They could be used at temperatures up to 870 and 900°C (1598 and 1652°F) respectively while Nimonic 95 allowed short-time service

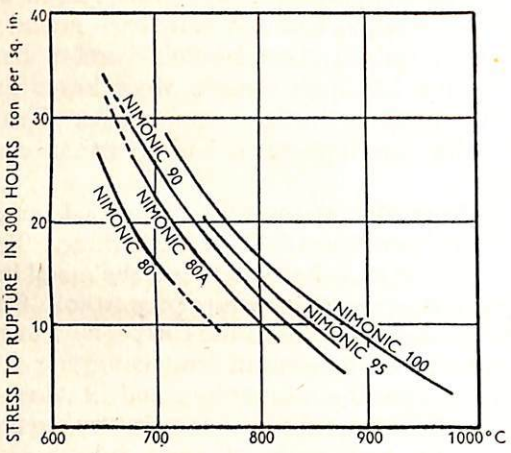


Fig. 1.—Comparison of the Nimonic alloys for 300-hour stress to rupture

at 925°C (1697°F). Forging of these alloys took place in the 1050–1150°C (1922–2102°F) region and the comparatively narrow range available meant that frequent reheating was essential. Consequently production costs tended to increase compared with earlier materials which were less difficult to work and this later led to the development of extrusion and precision casting techniques as a means of facilitating production.

Recently the production of a further alloy in this series was announced, namely Nimonic 100, and Figure 1 compares the various Nimonic alloys.

Fe-Ni-Cr-Co Alloys

The continued development of valve steels over the years had resulted in the production of an alloy designated G.2 having the composition

%C	%Si	%Mn	%Ni	%Cr	%W	%Mo	%Fe
0·5	1·3	0·7	13·2	13	2·5	0·1	balance

This was an austenitic steel and in 1943 work was in progress to further improve its properties. It was known that the addition of titanium or aluminium conferred superior creep strength on these steels and in the search for materials suitable for aircraft gas turbines attention was directed upon this alloy.

The early stages of the development of G.2, in which the carbide formers were added to the original alloy, were accompanied by difficulties associated with the low ductility arising from grain boundary carbide precipitation. Additions of columbium, molybdenum, titanium and vanadium were tried separately or in combination, and cobalt was also found to be useful in improving the creep strength. Ultimately G.18B was developed with the composition:

%C	%Si	%Mn	%Ni	%Cr	%W	%Mo	%Co	%Cb	%Fe
0·4	1·0	0·8	13	13	2·5	2·0	10	3·0	balance

and its original application was for turbine discs. It was, however, used for turbine blading where good creep strength in the 600–850°C (1112–1562°F) region was required. One of its merits was ease of welding by any of several methods and this rendered it particularly suitable for constructions in which the disc had to be welded to the turbine shaft. Several large forgings were later made from it, notably in the Sulzer Weinfelden plant where G.18B was used for both the turbine blades and the disc. The creep properties of this alloy were enhanced by warm working in the range 650–800°C (1112–1562°F) after a solution treatment at 1300°C (2372°F).

A variant, G.19, was produced with greater emphasis on scaling resistance and intended for casting. It could be used up to 950°C (1742°F) and was thus suitable for turbine nozzle blades and cast parts generally. In 1948 another alloy, G.32, which stemmed from the same origin, came into use for blades and discs. The cobalt content of this alloy had been increased to 46% and a casting alloy of similar composition (G.34) also became available although the use of this alloy was restricted to nozzle blades. Since all these alloys were high in the strategic elements an attempt was made to maintain the same properties but without the use of cobalt and columbium, and G.38 was produced to this end.

Early Ferritic Gas Turbine Steels

Because of their low creep strength at temperatures above 600°C (1112°F) the ferritic steels were, in the early gas turbines, limited to use in the comparatively cool components. As cooling techniques improved it was found possible to maintain the operating temperature of the turbine discs below this temperature in certain cases and this stimulated interest in the development of better ferritic steels. There was available a fairly wide range of these steels which had found application in, for instance, the steam turbine and these steels offered undoubted advantages in the construction and manufacture of turbine discs if their strength could be improved. The gas temperature in the first turbine designs was sufficiently low to allow the Cr-Si-Mo and Cr-Mo types to be used. As the power output of subsequent engines increased so the operating temperature rose and it became necessary to turn to the 3% Cr-Mo-V steels. Later the complex ferritic alloys were developed, in which chromium, tungsten, molybdenum, vanadium and columbium were used to produce a high creep strength at temperatures approaching 650°C (1202°F).

An illustration of this trend can be seen in the development of the Jessop-Saville range of ferritic gas turbine steels. An alloy known as H3 (6% Cr-Si-Mo) had originally been used for valves and was tried in the early gas turbines. Its creep strength, however, was low and it was soon replaced by H31, an example of the 1% Cr-Mo steels. The need to withstand still higher stresses was met by the introduction in 1945 of H40, a 3% Cr-Mo-V-W steel in which outstanding creep strength was available up to about 600°C (1112°F) as well as fairly good oxidation resistance. Its scaling properties were, however, not as good as was desirable and H46 was introduced in 1948. This alloy had a high chromium content (12%) with additions of molybdenum, vanadium and columbium, and its strength was comparable with the simpler austenitic steels at tem-

peratures of the order of 650°C (1202°F). A characteristic of these ferritic alloys is that their strength is developed by hardening and tempering, with the tempering taking place above the operating temperature in order to prevent softening in service.

TABLE II

Creep strength of some ferritic gas turbine steels

	STRESS TO GIVE 0·1% STRAIN IN 300 HOURS-TONS/IN^2		
	550°C	600°C	650°C
H3A ..	5·3	2·8	—
H31 ..	13	6·3	5
H40 ..	21	14	8
H46 ..	—	14	10·2

U.S.A.

The development of the exhaust gas turbo-supercharger in the U.S.A. which began during the First World War produced an early demand for high-temperature materials. Operating temperatures were somewhat higher than the 400–450°C (750–850°F) which was the limit in steam turbines at the time and a better material than the 12% chromium steel then in use for steam turbine blades was sought. Before 1928 two materials had been tried for turbo-supercharger blades operating at 600°C (1112°F). These were SAE 6150 and Silchrome No. 1, both low-alloy steels:

	% C	% Si	% Cr	% V	% Fe
SAE 6150 ..	0·5	0·3	1·0	0·2	balance
Silchrome No. 1 ..	0·5	3·0	8·5	—	balance

The first of these had originally been developed as a spring steel and Silchrome No. 1 as an exhaust valve steel. Both these uses required a good hot strength but apart from a superior corrosion resistance neither steel was much better than the ordinary constructional steels.

An improved material became available about 1930 when KE.965, which had been developed in England for exhaust valves, was adopted in the U.S.A. as a blade material. This had been designed for a good hot strength and allowed operating temperatures up to 760°C (1400°F):

% C	% Cr	% Ni	% W	% Fe
0·4	14	14	3	balance

A modification was produced by the Universal Cyclops Steel Corp. which replaced KE.965. This was 17W and differed from the English alloy in that the nickel content was greater and molybdenum was also added:

% C	% Cr	% Ni	% W	% Mo	% Fe
0·4	14	19	2·5	0·5	balance

By the middle 1930s more complex alloys had been developed and interest centred around three main groups of materials:

1. 18-8 type austenitic steels.
2. Low carbon-ferritic chromium irons with low nickel content.
3. Nickel-cobalt-iron-titanium alloys with additions of chromium or molybdenum.

An outstanding alloy of the third type was K.42B developed by the Westinghouse Company:

% Cr	% Ni	% Co	% Fe	% Ti
19	46	25	7·5	2·5

This was an age-hardening alloy possessing ample oxidation resistance as well as good high-temperature strength, and was used in the early Westinghouse gas turbines.

Up to about 1930 the best high-temperature materials had been based on iron with small amounts of alloying elements. The trend, however, was towards the use of increasing amounts of nickel and chromium in order to give better creep strength. At the same time as the high-nickel–high-chromium alloys came into general use a new type of high-temperature material emerged, based on cobalt instead of iron, and produced by the Haynes Stellite Company.

Although the use of these latter alloys as materials for high-temperature components dates from about 1930 the first of the series had been developed before 1914. Hardness and good hot strength were essential characteristics of these alloys, and they also possessed some degree of corrosion and oxidation resistance. Originally used in cutting tools, where toughness and hardness at high temperatures were needed, their usefulness in other high-temperature applications became evident about 1930 when the properties of Stellite alloy No. 6 were established. This material was extremely difficult to forge but it could be cast and the lost wax process was used to produce a set of turbo-supercharger blades in

Vitallium, a modification of Stellite No. 6 alloy used in dentistry. The casting process, however, was not particularly successful when first used for the manufacture of turbo-supercharger blades and a large proportion of defective blades resulted. The No. 6 alloy was very hard owing to the high carbon content, and consequently of rather low ductility. It was the general opinion therefore, at this time, that these cast blades would be too brittle and it was not until 1941 that a Stellite alloy appeared which was suitable for casting these components. Soon after the commencement of the Second World War it was obvious that the need to increase the production of turbo-supercharger blades would impose a serious strain on the forging capacity in the U.S.A. Thoughts had therefore turned again to the production of cast blades. As already mentioned, when they were first introduced the cobalt base alloys were hard and possessed a tendency to brittleness, but since the carbon content largely influenced the hardness, medium carbon alloys were rapidly developed.

By 1941, therefore, a suitable casting alloy known as Stellite No. 21 was available. This material differed from Vitallium in that a small percentage of nickel was added, while to reduce hardness the carbon content had also been lowered from 0·5 to 0·25%. The hardness of this material, which had originally been about 55-60 Rockwell C, was thus reduced to about 48 Rockwell C, and it was further reduced to 40-42 Rockwell C with the development of a controlled production process. This material was used most successfully for the blades of the GE type B turbo-supercharger, in which they were welded to the disc by the Unionmelt process.

By the time gas turbine production in the U.S.A. had developed momentum in 1944 a well-established material for cast blades was thus already available. This material was used for the blades of both the GE.I.16 turbo-jet engine and the axial turbo-prop TG.100. By increasing the nickel content still further Stellite No. 21 was modified to give Stellite No. 23 which possessed better high-temperature strength without the disadvantage of undue room-temperature brittleness.

The Nickel-molybdenum Alloys

These alloys were originally developed for resistance to attack by hydrochloric acid and their properties were generally established by about 1915, although extensive development work was necessary before they became available commercially in 1926. It was during this period of development that their outstanding high-temperature strength first became apparent.

Three alloys were produced under the names Hastelloy A, C

and D, the first in wrought form and the last two as casting alloys. Hastelloy B, containing more molybdenum (28%) was developed rather later than the others.

The G.E.C. turbo-supercharger provided an early application for materials of this type in the high-temperature field and much associated development work was carried out. On the basis of short-time tests at 800°C (1470°F) the cast forms of the A, B and C alloys showed superior stress-rupture properties. Below about 750°C (1380°F) the wrought form of Hastelloy B had very good all-round properties but since it contained no chromium its oxidation resistance was inadequate for use above this temperature. Hastelloy C, however, did contain a proportion of chromium and could be used at 800°C (1470°F).

All these alloys were far from easy to hot work. Carefully controlled forging with a small reduction in size at each heat was essential, while the life of the forging dies was very short. A further feature of this type of alloy was the very close control over composition required.

Wartime Research and Development

The immensely accelerated programme of high-temperature material research and development which was undertaken in the U.S.A. after 1941 owed its impetus mainly to the urgent need to produce alloys suitable for use in the aircraft gas turbine. Although the aircraft turbo-supercharger driven by exhaust gases had provided a demand for heat-resisting alloys which originated during the First World War, this particular field of application was too small and not sufficiently important to call for the concentrated metallurgical effort later initiated by the advent of jet propulsion. By 1939 several of the aircraft corporations had begun to consider the initial project design of a jet engine; the Durand Committee on Jet Propulsion came into existence in 1941 and development contracts were awarded to Allis-Chalmers, General Electric and Westinghouse. Meanwhile the Whittle engine had been sent to the U.S.A. to be manufactured under licence. The sudden and urgent need for blade and disc materials led, therefore, to the initiation in 1941 of two major research projects[1] which were designed to cover an extremely wide range of possible high-temperature alloys.

In the first of these projects the emphasis was on the ability of the material to carry high stresses for long periods at temperatures

[1] Project N.R.C.8.—carried out under the auspices of the War Metallurgical Committee and the National Research Council. N.A.C.A. Project—controlled by the Sub-committee on Heat-resisting Alloys.

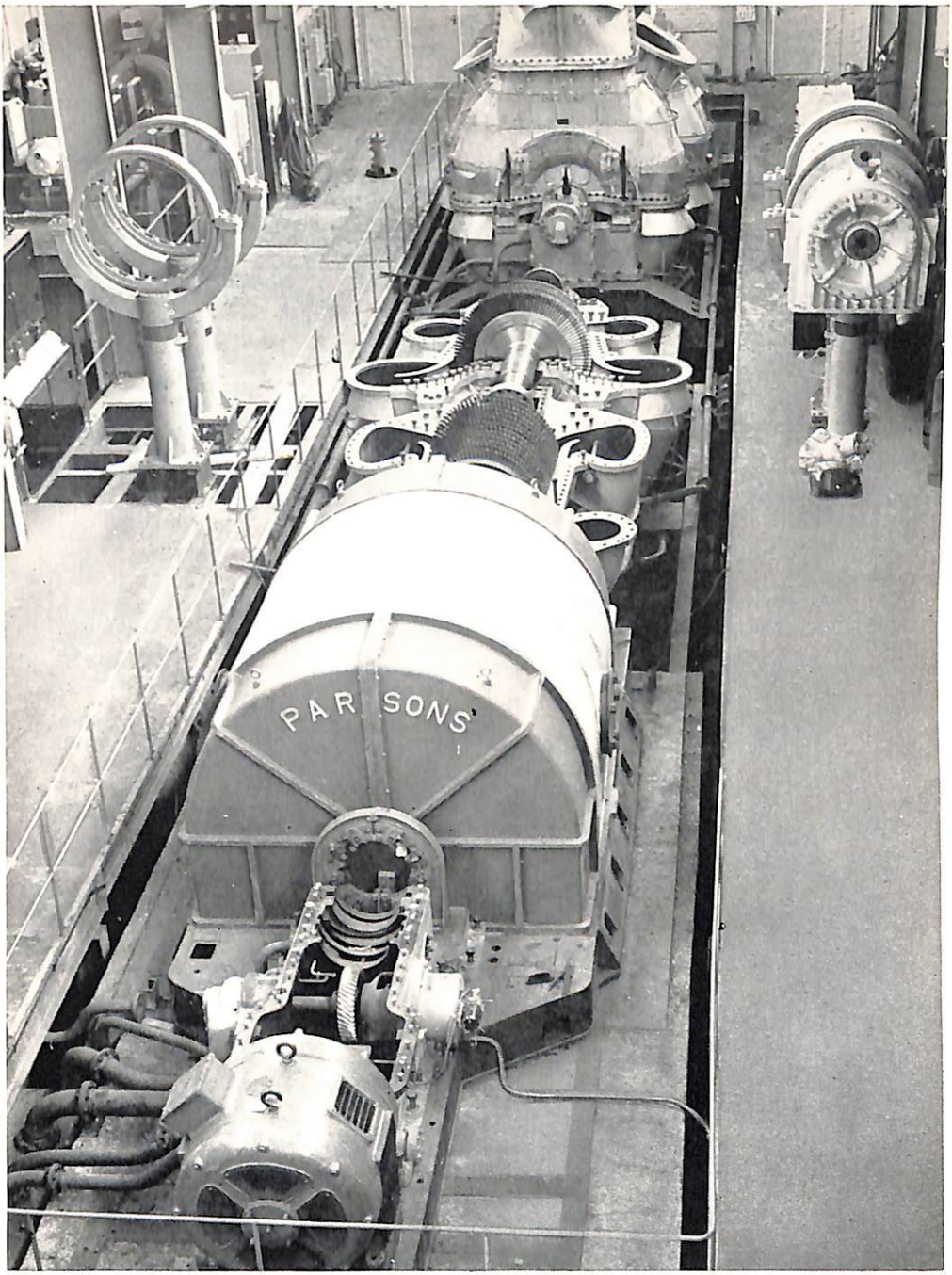


Plate 3. C. A. Parsons and Co. Ltd. 10,000-kW double compound gas turbo-generator in course of installation at the National Gas Turbine Establishment, Pyestock

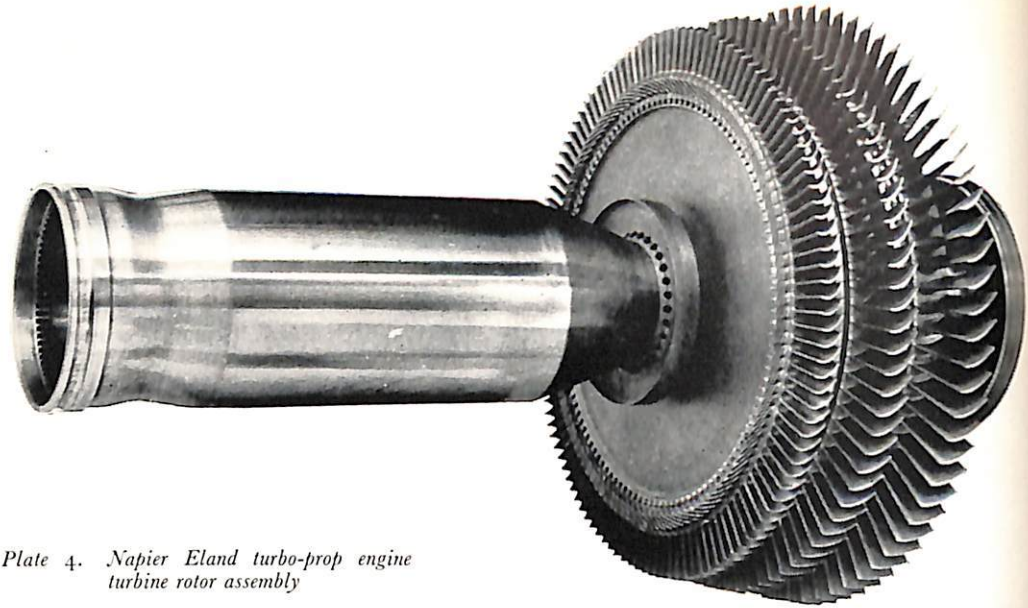
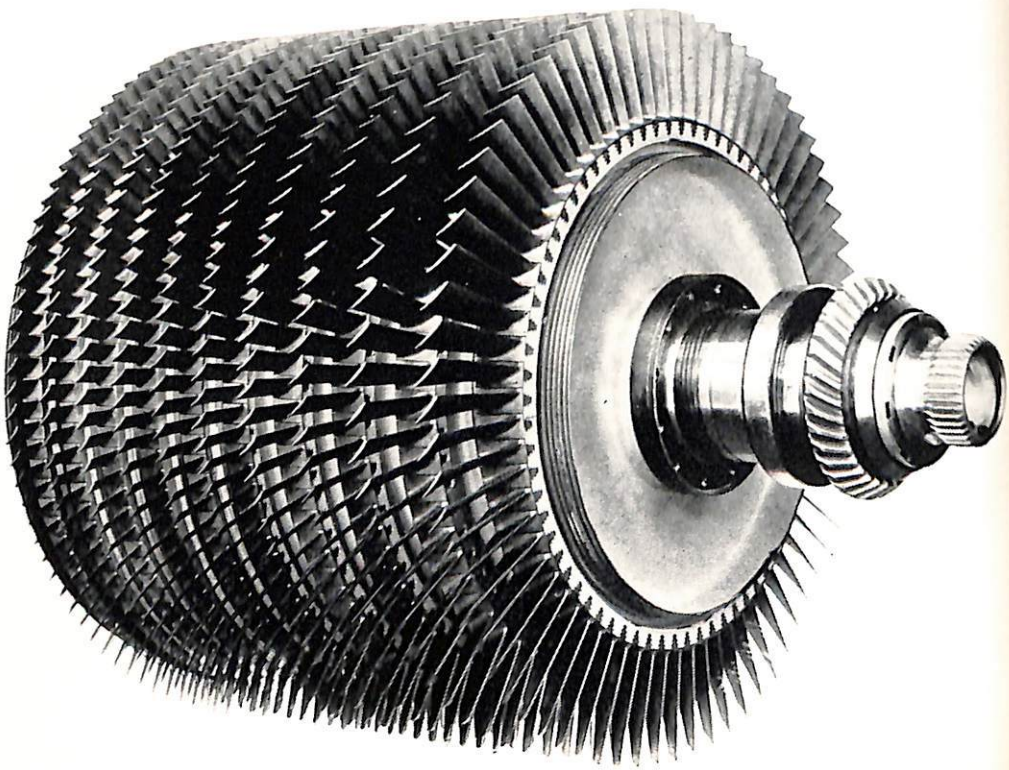


Plate 4. Napier Eland turbo-prop engine turbine rotor assembly

Plate 5. Napier Eland turbo-prop engine compressor rotor assembly



generally in the region of 815°C (1500°F) but in some cases at 735°C (1350°F). Some 86 austenitic-type alloys were examined, in the main of the Cr-Ni-Fe, Cr-Ni-Co-Fe, Co-Cr and Co-Cr-Ni types. The physical properties assessed were short-time tensile strength, creep and rupture strength, while an attempt was made to indicate the relative suitability for high-temperature applications of the materials considered. From these investigations the cast cobalt-based alloys emerged as outstanding although initially there was a certain reserve in accepting their suitability as a turbine blade material.

The second project emphasised a short life use of under 1000 hours and was designed to select alloys having the best combination of yield strength both at room temperature and at 650°C (1200°F). Only five alloys out of the 84 tested were sufficiently promising to be considered for potential use in the gas turbine. One of these was the already well-established Timken alloy 16% Cr-25% Ni-6% Mo which subsequently found extensive application as a turbine disc material. One purpose of the investigation was to examine the relationship between heat treatment, fabrication methods and high-temperature strength but no systematic correlation could be found.

The first of the General Electric I series gas turbines, which was based on the Whittle W.1X and W.2B engines, employed American materials already available in 1939. Rex 78, which had been used for the turbine blades of the W.1X engine, was replaced by Hastelloy B, and both Vitallium and Nimonic 80 were also tried. Turbine rotor blade failures were much fewer in this engine than in the original Whittle engine since the high-temperature strength of the American alloys was superior to Rex 78, the only suitable blade alloy available at the time of the W.2B design. One notable change was the alternative of precision-cast rotor blades introduced as a consequence of an acute shortage of blade forging capacity. A considerable number of the I series engines had cast turbine blades and by 1945 a completely satisfactory production technique had been evolved.

TABLE III

General Electric I series engines—turbine nozzle inlet 802°C (1485°F)

SOME HIGH-TEMPERATURE MATERIALS USED IN CONSTRUCTION	
Turbine disc	17W (later replaced by Timken 16-25-6)
Turbine rotor blades	Vitallium, 19-9DL, Hastelloy B (forged) or precision-cast Stellite 21
Turbine Nozzles	18-8 steel, precision-cast Stellite 21
Sheet Metal Parts	Inconel (80% Ni-14% Cr-6% Fe) or 18-8 stainless steel.

In the early Westinghouse engines which, in contrast to the General Electric engines, were of purely American design, the existing Westinghouse K.42B alloy was used with some success for the turbine blades, together with Vitallium for the nozzles and a 19% Ni-9% Cr stainless steel for the disc. Quantity production difficulties led to the use of cast turbine blades, for which Stellite 23 was used, a similar alloy to Stellite 21 and also cobalt based.

Germany

The systematic development of heat-resisting alloys commenced in Germany about 1925. As in America the exhaust gas turbine supercharger was undergoing development and provided the incentive in the search for superior materials. Investigations were in progress at the D.V.L. Institute into the complex alloys for use at temperatures approaching 800°C (1470°F) and the industrial organisations of B.M.W., Junkers, Krupp and Böhler had begun development of austenitic alloys intended for blades and rotors.

The initial stages of the D.V.L. research consisted of a systematic series of creep tests on known alloys thought to have good high-temperature properties. These were necessarily short-time tests and creep strengths were quoted on the basis of the stress to give 0·1% strain in 300 hours. This was known as the D.V.L. strength. Since the investigations were concerned with materials for aircraft use the short-life tests provided a reasonable criterion. The results of these investigations were published in 1938 and indicated the relative merits of a wide range of austenitic steels. These fell into three main groups:

1. Ni-Cr steels with W, Mo, Va, Ti and Ta additions.
2. Fe-Ni-Cr-Co complex alloys with W, Mo, Ti or Ta.
3. Cr-Mn steels.

Cobalt-base alloys had been tested at the D.V.L. Institute before 1938 and it had been shown that up to 735°C (1350°F) they were superior to any other blade material. Casting cobalt-base alloys of the Vitallium type were also well known. These alloys were not developed at D.V.L. despite the knowledge of their merit, due mainly to the realisation that since cobalt was a strategic element it would be restricted in the event of war.

Among the materials tested by D.V.L. in 1935–36 was a Krupps steel P.193. This was a Ni-Cr austenitic steel with the addition of titanium. It possessed good corrosion resistance and a satisfactory hot strength which was produced by solution heat treatment followed by precipitation hardening. Research at Krupps produced

by 1937 a superior version called Tinidur which could be used up to 750°C (1380°F). The composition eventually developed was:

% C	% Si	% Mn	% Ni	% Cr	% Ti	% Al	% Fe
0·15	0·5	0·8	30	15	1·7	0·2	balance

In the development of this material a number of other carbide formers were tried but the best results were given by the use of titanium. The effect of replacing the varying proportions of iron by nickel was also studied together with cobalt additions up to 30% but these changes were not effective in markedly improving the creep resistance. Initially the alloy could only be hot worked but reduction of the carbon content to less than 0·15% removed this disability and allowed cold working. To some extent Tinidur suffered from embrittlement when exposed to prolonged high-temperature heating and the precipitation hardening effect fell off at about 700–750°C (1300–1380°F).

With the advent of the German aircraft gas turbines Tinidur was selected for the turbine rotor blades. Some trouble was experienced in controlling the properties of the finished components for considerable differences in service life were found in blades which appeared to be made from identical materials. This same trouble in controlling the properties of the finished article had been experienced both in the U.S.A. and Britain until the importance of the precise control over both the composition and the production was realised and achieved. The development of heat-resisting materials in Germany was greatly influenced by the potential shortage of the alloying elements nickel, cobalt and chromium. At the outbreak of war the potential shortage became actual and nickel became steadily scarcer; this resulted in another Krupp alloy, Chromadur, replacing Tinidur. In this alloy the nickel was entirely replaced by manganese as the austenite-forming agent:

% C	% Si	% Mn	% Cr	% V
0·1	0·7	18	12	0·7

and although Chromadur had better welding properties, an advantage in the hollow-blade construction adopted by the Germans, the creep strength was inferior to Tinidur.

Difficulties associated with the production of the highly-alloyed austenitic materials led to the use of ferritic steels to a steadily increasing extent as the war progressed. There was a variety of types, some of which had previously been used in the steam turbine, but which could not be used in the gas turbine without some form of cooling since their creep strength was very low at temperatures

above 550°C (1020°F). In general these fell into the following groups, in which the strength is more or less in ascending order.

1. Mn-V.
2. Cr-Mo.
3. Cr-V.
4. Cr-Mo-V.
5. Cr-Mo-V-W.

The operating temperatures for which these steels were suitable ranged from about 400°C (750°F) in the case of the Mn-V steels to about 550°C (1020°F) for the more highly alloyed steels with higher chromium. Eventually the acute molybdenum shortage restricted the use of some of these alloys but not before the Cr-Mo-V-W type had been found to give very good results when used for turbine rotors in the aircraft gas turbines.

Post-war Development in the High-temperature Alloy Field

It will be clear from the foregoing that, in the early days of the gas turbine, use was made of materials primarily developed for quite different purposes. Since this period was quickly followed by the war years, when tremendous impetus was given to the application of the gas turbine to aircraft propulsion, it was not until the post-war period that serious attention was paid to the material requirements of gas turbines for industrial use. Alloys known to be suitable for aircraft use, where engine life was restricted to about 300 hours, were comparatively unknown quantities when engine lives of 100,000 hours were called for.

While research and development connected with the gas turbine has continued, therefore, over the last decade or so and still goes on, many of the wartime alloys have passed into history and only a dozen or so alloys remain as established gas turbine materials. Research has tended to concentrate on determining and improving the behaviour of the more promising blading alloys in tests lasting many thousands of hours and since this of necessity calls for much time and expense, the production of quite new alloys has greatly diminished. It is also quite likely that the alloys now available represent almost the best that can be expected from the three metals, iron, nickel or cobalt. For really outstanding advances in material development, attention is turning to the more refractory metals such as chromium and molybdenum, ceramics and metal-ceramic mixtures. All these present many very difficult problems and the present state of development in these fields will be dealt with in later chapters.

REFERENCES

1. J. Riley. *J. Iron St. Inst.*, 1889, p. 45.
2. C. Bach. *Ver. deutsche Ing.*, 1903, p. 762.
3. J. H. S. Dickenson. *J. Iron St. Inst.*, 1922, Vol. 106, p. 103.
4. P. Chevenard. *Comptes Rendus Academie Des Sciences*, 1919, Vol. 169, p. 712.
5. F. C. Lea. *Proc. Instn. Mech. Engrs., Lond.*, 1924, Vol. 2, p. 1053.
6. H. J. French and W. A. Tucker. *Proc. Amer. Soc. Test. Mater.*, 1924, Vol. 24, p. 56.
7. W. Rosenhain and C. H. M. Jenkins. *J. Iron St. Inst.*, 1930 (1), p. 225.
8. Iron and Steel Institute. *Proc. of the Symposium on High-temperature Steels and Alloys for Gas Turbines.*
9. C. Sykes. *J. Iron St. Inst.*, July 1947, p. 156. "Steels for Use at Elevated Temperatures."
10. B.I.O.S. Report No. 396. "Report on Visit to Germany and Austria to Investigate Alloys for Use at High Temperature."
11. A.S.T.M. *Symposium on Materials for Gas Turbines*, 1946.
12. R. O. Schlaifer and S. D. Heron. *Development of Aircraft Engines and Fuels.*
13. W. T. Griffiths. "The Problem of High-temperature Alloys for Gas Turbines", *J. R. Aero. Soc.* 52, Jan. 1948, pp. 1–26.
14. F. Whittle. "The Early History of the Whittle Jet Propulsion Gas Turbine", *Proc. Instn. Mech. Engrs., Lond.*, 1945, Vol. 152, p. 419.

TWO

(Section One)

The Influence of Performance Requirements on Operating Conditions of the Gas Turbine

Much of the need for development of new materials with strength at high temperatures arises from the manner in which temperature and pressure affect the performance of the gas turbine cycle. Both thermal efficiency and specific power output, which are major performance criteria, are influenced by the maximum cycle temperature and the pressure ratio. Broadly speaking the maximum cycle temperature has the more important effect on specific output whereas the pressure ratio has the greater influence on the thermal efficiency. Because of the numerous forms the gas turbine may take, in practice there is not a unique relationship between these two parameters and the performance, and each design needs to be analysed in detail, taking into account component efficiencies and pressure losses, in order to arrive at a precise knowledge of the way in which maximum temperature and overall pressure ratio affect performance. In general terms, however, a very useful impression may be obtained by considering an idealised version of the gas turbine cycle in which there are no pressure losses and in which component efficiencies of 100% are postulated. Although the conclusions reached from studying such a cycle would not be directly applicable to actual engines, the discrepancies are likely to be of a quantitative nature only and the overall picture otherwise remains much the same. Since the ideal cycle may be considered as that which designers endeavour to achieve in practice the consideration of the generalised ideal case is justified. For understanding the operating conditions chosen by the designer and how these may influence the choice of materials it can be more useful than a variety of individual studies of actual engines.

Most of the gas turbines which have been manufactured to date have been of the open-cycle type in which the air is drawn into the engine and rejected, a fresh charge continuously replacing that rejected. The other version of importance is the closed cycle in which the same working fluid is continuously recirculated. This is based on a different thermodynamic process in which the importance of high-temperature levels is not as great as in the open-cycle gas turbine. By raising the minimum cycle pressure, as is possible

in a closed circuit, specific powers may be obtained with the use of lower maximum temperatures than would be needed in an open cycle design. This version of the gas turbine is, however, more complicated, since it involves heat exchangers, and is at present considered suitable only for heavy land installations.

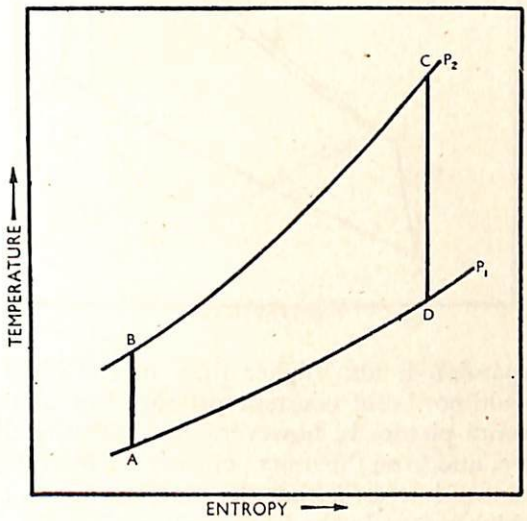


Fig. 2. — Temperature-Entropy diagram for the ideal constant pressure cycle

The Brayton or constant-pressure cycle is the basis of the ideal thermodynamic processes in the open-cycle gas turbine, and is shown diagrammatically in Figure 2. It postulates

1. Isentropic compression from A to B.
2. Heat addition at constant pressure from B to C.
3. Isentropic expansion from C to D.
4. Heat rejection at constant pressure from D to A.

The actual cycle would not be so very different from this. In practice it would not be possible to achieve either isentropic compression or isentropic expansion although with the component efficiencies now possible the departure from isentropic processes is not unduly great. Furthermore the process of heat addition would in practice be accompanied by a pressure loss. Heat addition involves a pressure loss due to the rise in temperature from B to C as well as a loss due to flow of the gases through the combustion chamber. The actual cycle would resemble Figure 3 in which the temperature at the end of compression is higher than in the ideal case for the same pressure ratio, and the temperature at the end of

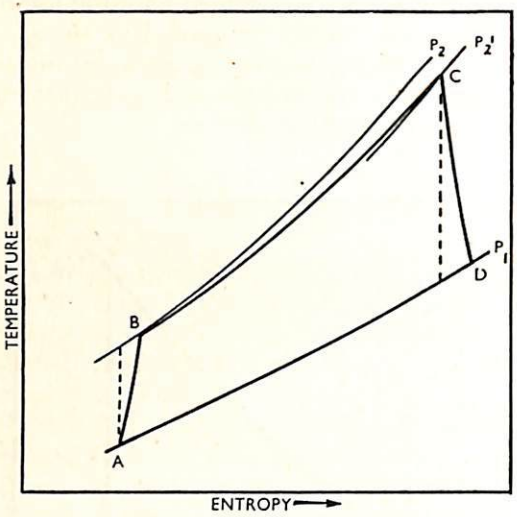


Fig. 3. — Temperature-Entropy diagram for the practical cycle

expansion is also higher than in the ideal case. Heat addition would not be at constant pressure but along some line BC. The overall picture is, however, not markedly different from the ideal case, and from the aspect of material operating conditions the main point of interest is that the temperature at the end of compression is higher than in the isentropic case.

In the actual gas turbine the cycle processes of Figure 2 are carried out stage by stage in (a) a compressor, (b) a combustion chamber in which fuel is burned to produce the temperature rise, and (c) the turbine proper. The manner in which the components are linked will depend on the particular application. It would for example be possible to carry out the compression from A to B in one or more compressors followed by a corresponding splitting-up of the expansion between one or more turbines: the basic open cycle is capable of assuming a number of different forms.

The simplest form is shown in Figure 5 where the compressor is driven by the turbine in which only part of the expansion is completed, the remainder of the expansion is taking place in the exhaust nozzle or jet pipe. This is the form used in the turbo-jet engine. From this it is possible to see the general level of temperatures and pressures which prevail in the different components when a choice of pressure ratio and maximum cycle temperature has been made for the cycle. A slightly more complicated version, the turbo-prop configuration, is shown in Figure 7 (*page* 30). Here there are two turbines, one driving the compressor, the other driving the propellor, while the remainder of the expansion takes place in

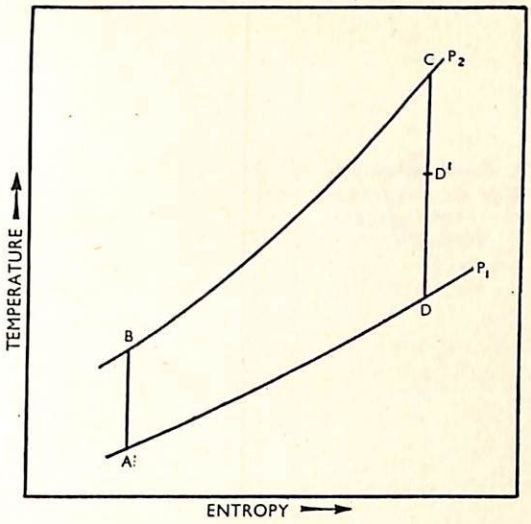


Fig. 4. — Temperature-Entropy diagram for the turbo-jet engine — ideal cycle

the jet pipe. A simple industrial set might use much the same combination as the turbo-prop except that all the expansion would be completed in the two turbines. Thus it may be observed that in the ideal constant-pressure cycle the pressure ratio determines the temperature at outlet from the compressor, for given entry conditions, and the maximum gas temperature exists at entry to the first turbine.

When the parameters which determine thermal efficiency and specific output in the ideal simple cycle are considered it can be easily demonstrated that thermal efficiency is dependent only on the pressure ratio while specific output is governed by both pressure ratio and maximum cycle temperature. The curves of Figures 8 and 9 show the relationship. In an actual engine when pressure losses and component efficiencies have to be considered these simple

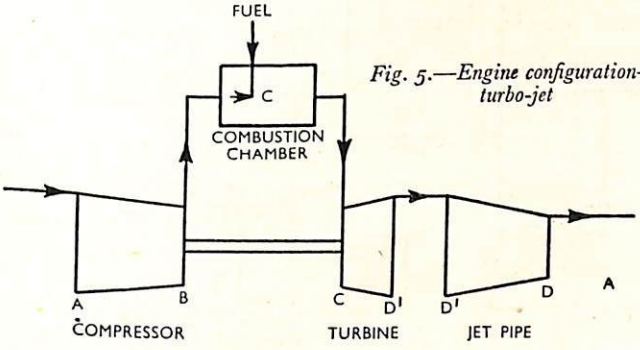


Fig. 5.—Engine configuration—turbo-jet

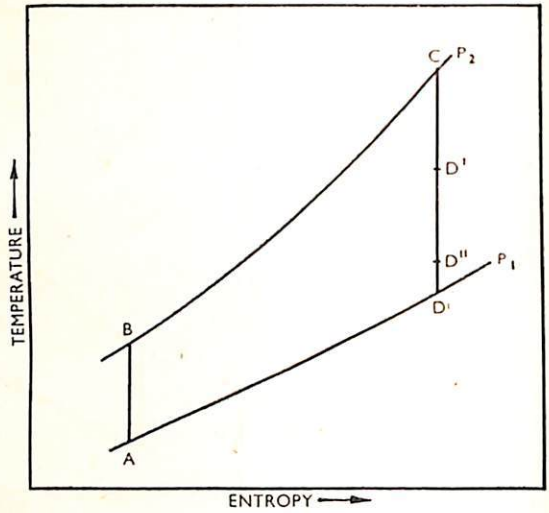


Fig. 6.—Temperature-Entropy diagram for the turbo-prop engine—ideal cycle

curves are modified and thermal efficiency becomes a function of both pressure ratio and maximum temperature; the general effect being not only to substantially reduce the efficiencies of Figure 8 but, for a particular T_{max}, to cause a decrease in efficiency after a certain pressure ratio is reached. Although increasing the pressure ratio up to about 10:1 produces increased efficiency, there is not the urge towards extremely high pressures in practice. The influence of cycle losses on the specific power is again to reduce the output considerably from the ideal case but the qualitative effect of maximum temperature and pressure ratio is much the same. The precise combination of pressure ratio and maximum temperature

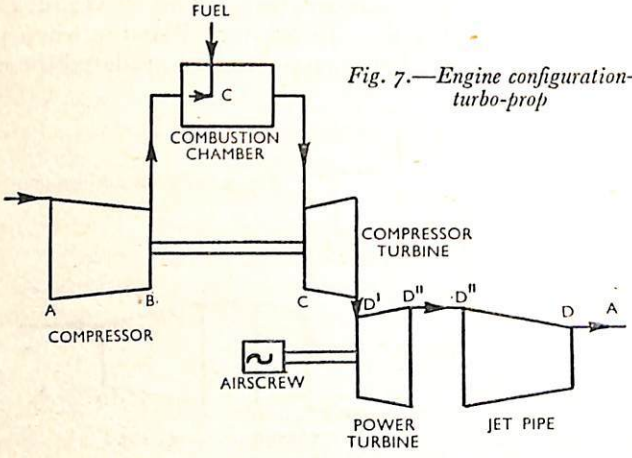


Fig. 7.—Engine configuration—turbo-prop

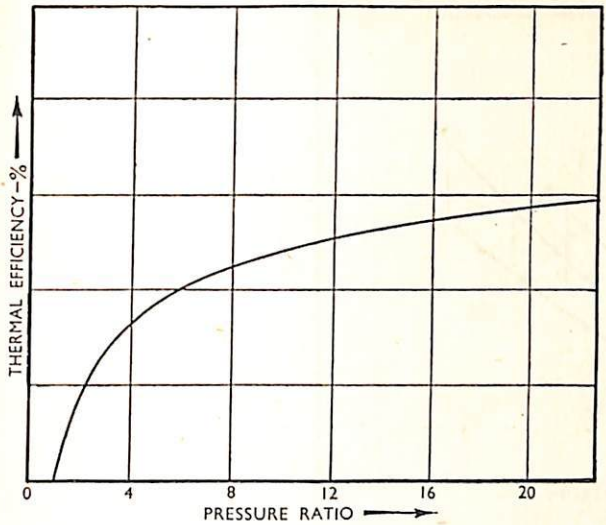


Fig. 8.—Relationship between the thermal efficiency and the pressure ratio in the simple ideal constant pressure cycle

used will always depend on the type of engine. In the air both a high efficiency and high specific power are obviously important, but in land installations efficiency is probably of much greater importance. However, these considerations do not alter the conclusion that it is in the interests of good performance in the simple open cycle to use both a moderately high pressure ratio and a high cycle top temperature.

There are a number of modifications to this simple version of the gas turbine cycle which can improve performance at the expense of

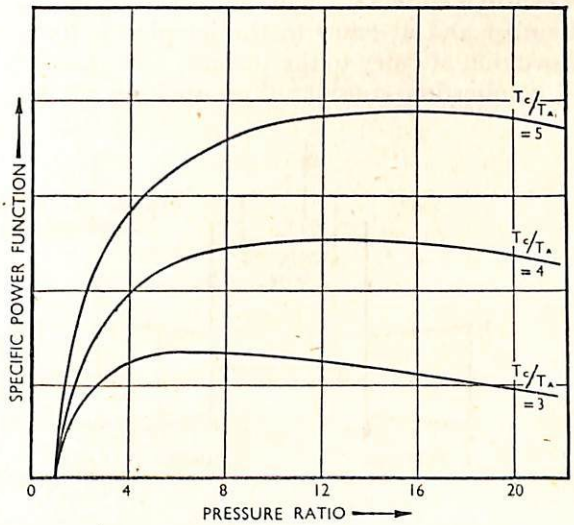


Fig. 9.—Relationship between the specific power function and the pressure ratio in the simple ideal constant pressure cycle

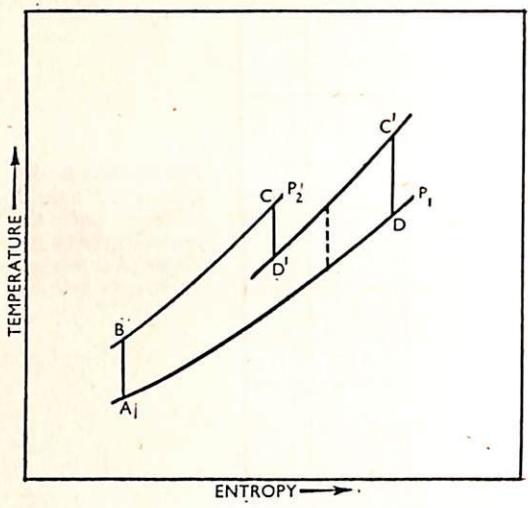


Fig. 10.— Temperature-Entropy diagram for the turbo-jet engine with reheat —ideal cycle

increasing complexity. Only one of these has been popular with aircraft gas turbine designers and that is the inclusion of reheat. The use of additional complications such as heat exchangers, compressor intercooling, and reheated turbine expansion, singly or in combination, is likely to be confined to land installations. Reheat in the aircraft gas turbine has a big effect on specific output for it is virtually the employment of high maximum cycle temperatures while exposing the turbine to only an intermediate level of temperature. The open cycle with reheat for a turbo-jet engine is shown in Figure 11. The temperature in the reheat combustion chamber and at entry to the jet pipe is likely to be considerably above that at entry to the turbine. In practice it would be limited by combustion considerations such as choking in the nozzle or

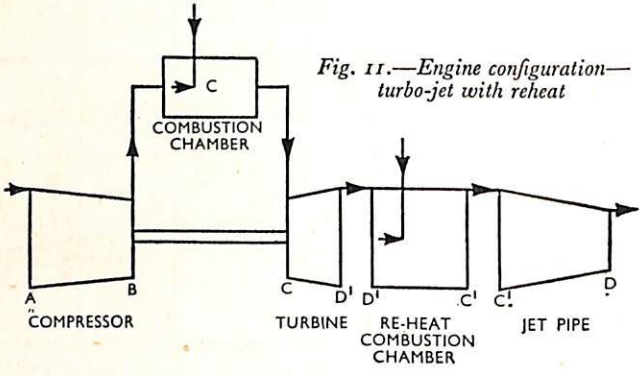


Fig. 11.—Engine configuration— turbo-jet with reheat

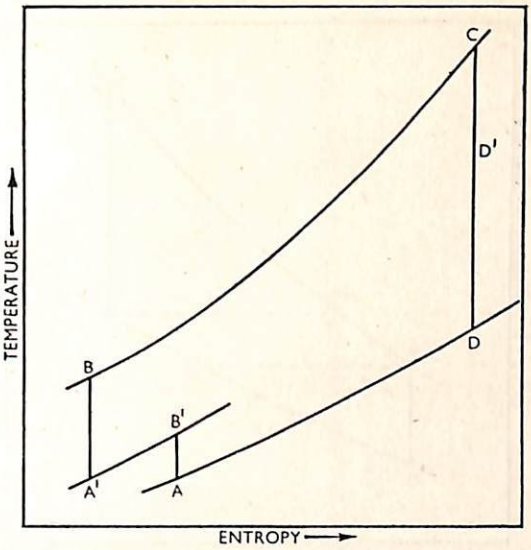


Fig. 12.—Temperature-Entropy diagram for the ideal constant pressure cycle with intercooling

maximum fuel flow limitations, but the higher the reheat temperature the greater the power output. In the industrial type of engine it would only be possible to reheat to the original turbine inlet temperature, which is fixed by the material limitations, for power must be developed at the shaft since it is not possible to use the jet energy.

Intercooling may be resorted to occasionally in order to reduce the compression work and improve specific output. This consists of cooling the gases at an intermediate stage of compression, as shown in Figure 13.

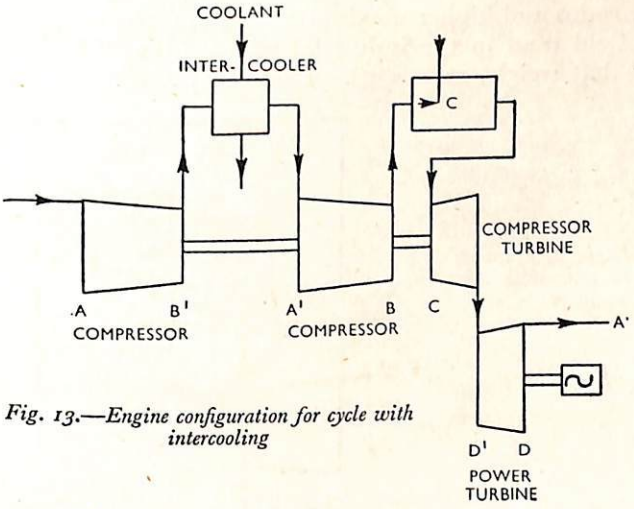


Fig. 13.—Engine configuration for cycle with intercooling

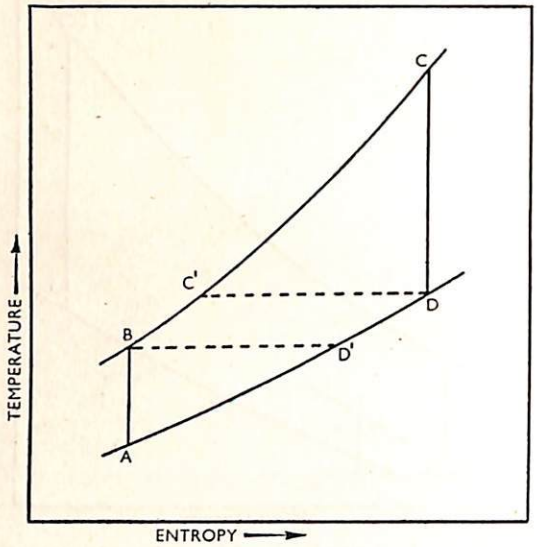


Fig. 14.—Temperature-Entropy diagram for the ideal constant pressure cycle with heat exchange

Heat exchange illustrated in Figure 15 is an attempt to reduce the amount of heat in the rejected gases and so improve the cycle efficiency and part load performance. This is applicable when the gas temperature at the exit from the turbine is above the air temperature at the exit from the compressor. This modification is more effective at low pressure ratios and hence pressures, and consequently stresses, in the open-cycle heat exchanger are not usually high.

Consideration of actual gas turbine designs shows that higher pressure ratios and higher maximum temperatures are in use in the aircraft field than in the industrial gas turbine. The reasons for this are that weight must be kept low in the aircraft gas turbine

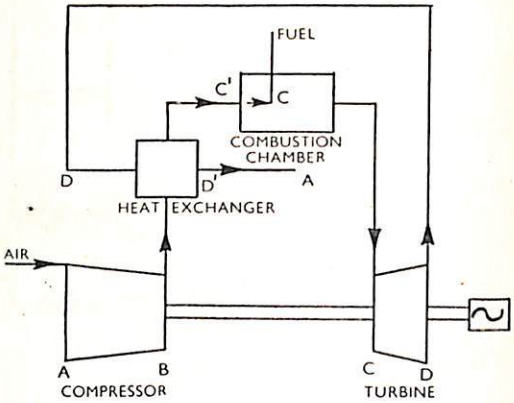


Fig. 15.—Engine configuration for the open cycle with heat exchange

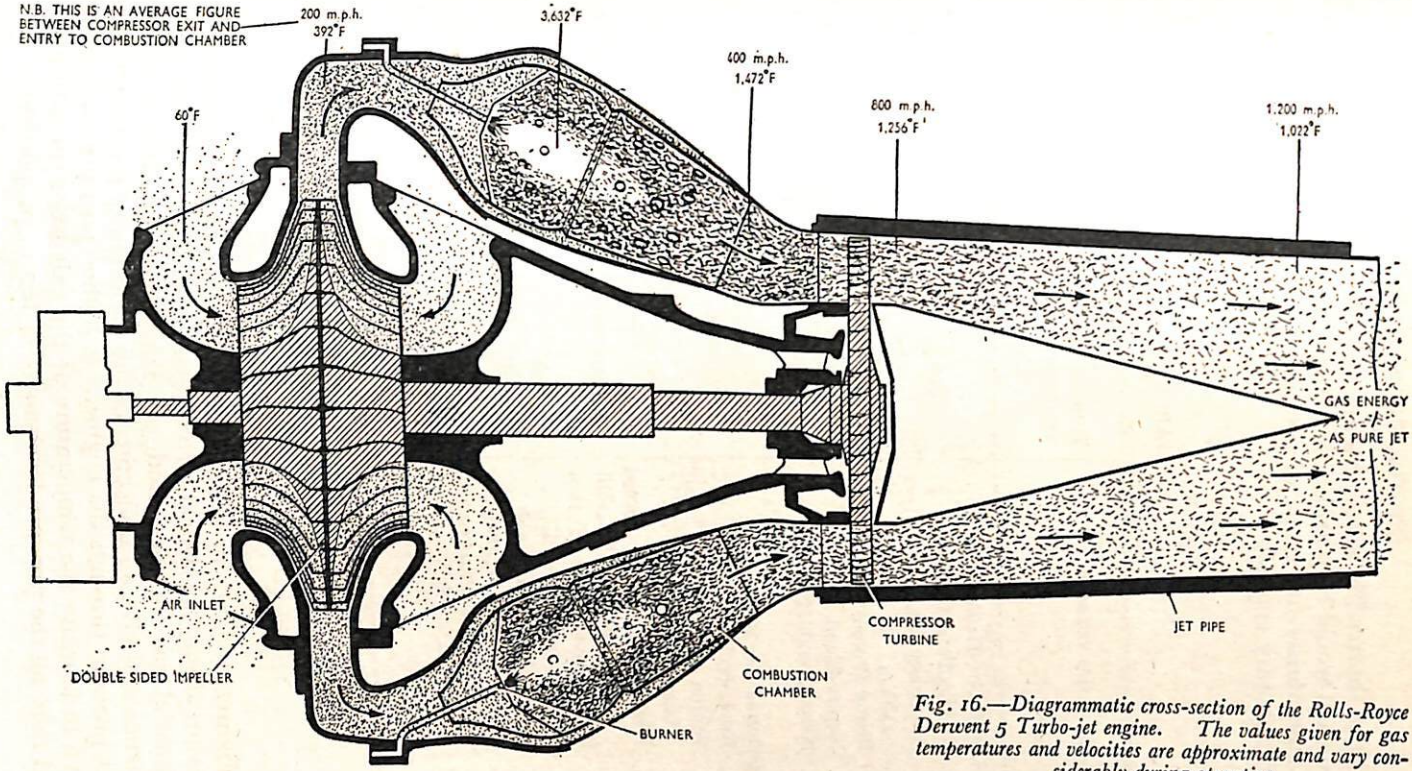


Fig. 16.—Diagrammatic cross-section of the Rolls-Royce Derwent 5 Turbo-jet engine. The values given for gas temperatures and velocities are approximate and vary considerably during operation

and the shorter life allows higher stresses and temperatures to be used than would be permissible in long-life industrial plant. Table IV summarises the range of pressures and temperatures in use in contemporary engines.

TABLE IV

GAS TURBINE	TURBINE INLET TEMPERATURE (°C)	TURBINE EXIT TEMPERATURE (°C)	PRESSURE RATIO
Aircraft Gas Turbines			
De Havilland Goblin 35 ..	800	650	3·67 : 1
Armstrong Siddeley Viper ASV.5	830	645	3·5 : 1
Bristol Proteus 755.. ..	845	500	7·2 : 1
Napier Eland. N.E1.1 ..	N.R.	N.R.	7·0 : 1
Avro Orenda. Series 10 ..	N.R.	720	5·5 : 1
Industrial Gas Turbines			
Rover 60 b.h.p.	790 max.	600 max.	2·9 : 1
Ruston and Hornsby Peat-burning turbine	725	430	3·6 : 1
John Brown and Co.'s Closed Cycle Turbine, Foleshill	600	360	4·0 : 1
B.T.H. 2500-kW Gas Turbo-alternators, Nairobi ..	650	400	4·5 : 1
English Electric EM.27P ..	777	490	4·8 : 1
C. A. Parsons 15,000-kW Gas turbo-generator, Dunston "A" power station	650	N.R.	8·0 : 1

N.R. = Not released.

Figures 16 and 17 show the distribution of temperature and pressure through a typical turbo-jet engine and for a medium-power industrial gas turbine. Both are fairly representative of current practice and give an indication of the variation of gas temperature and pressure through the engine. A higher pressure ratio would serve to increase the temperature at the exit from the compressor and raise all the pressure levels more or less in proportion.

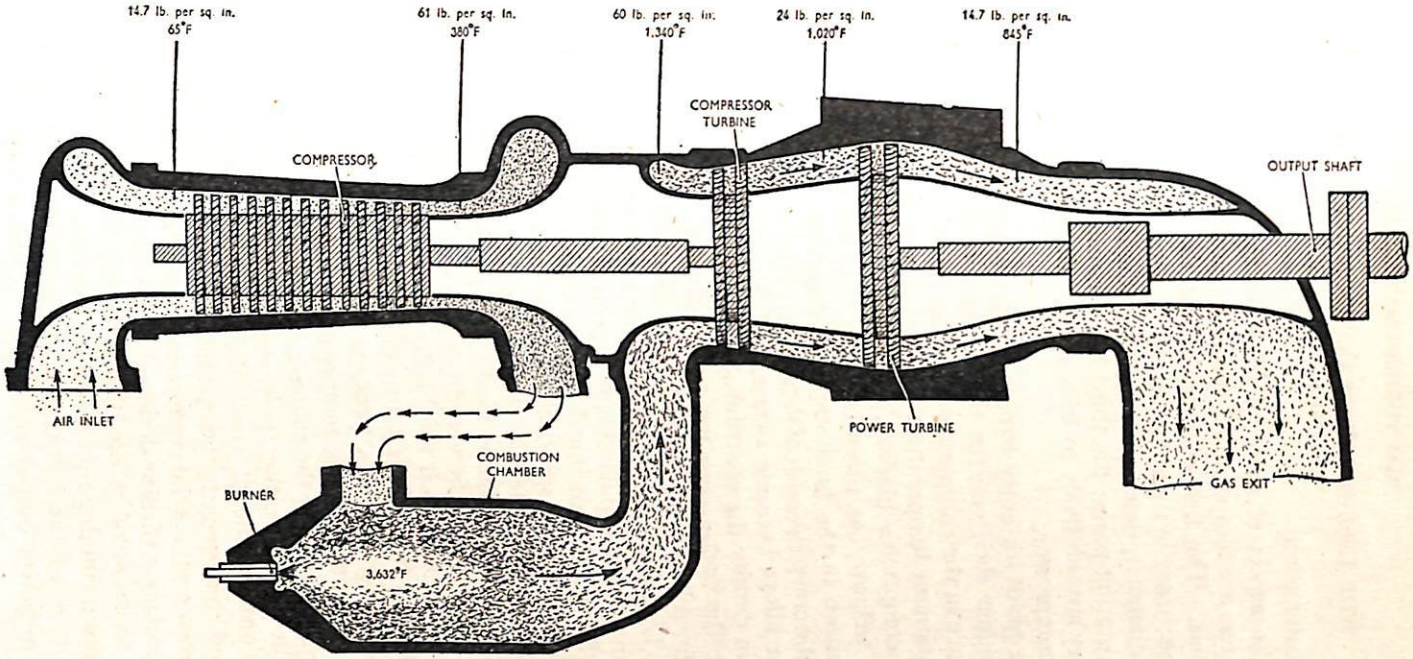


Fig. 17.—Diagrammatic cross-section of the Ruston and Hornsby Industrial Gas Turbine, showing the order of temperature and pressure in the various components

THE INFLUENCE OF COMPONENT DESIGN ON MATERIALS SELECTION

The Compressor

From the aspect of material selection the most important design parameters are temperature, speed of rotation and the aerodynamic conditions. The first of these affects the creep strength requirements necessary in the material; the second, in conjunction with the engine diameter, determines the centrifugal stresses which are set up in the rotating parts; the third has a considerable influence on the vibratory stresses likely to be set up in the blades during operation of the compressor.

For a given air entry temperature, the pressure ratio selected by the designer then governs the increase in temperature which will take place in the compressor. In the lower ranges of pressure ratio the maximum temperature reached allows the use of light alloys for the compressor blading, discs and casings if required for a low overall weight. At pressure ratios over about 6:1, however, the temperature in the final stages of compression reaches a level at which the creep strength of these alloys is insufficient and a change to stronger alloys becomes necessary. Since this also implies an increase in density the centrifugal forces increase at the same time.

Once the design pressure ratio is chosen, the rotational speed is determined largely by the size of the compressor. Where there is a need to keep the overall weight of the gas turbine as low as possible, i.e. in an aircraft application, the speed of rotation is correspondingly high. Since the centrifugal loading varies as DN^2, where D is the diameter and N the speed of rotation, increasing the speed of rotation has a considerable effect on increasing the stresses in the blading.

In a high aerodynamic blade loading design of compressor, the air velocities tend to be very high and this in turn imposes appreciable bending forces on the blades as the air direction is changed. The blade is thus subjected to vibrational forces from this source.

In selecting a design pressure ratio and rotational speed the engine designer fixes the temperature and pressure at the outlet of the compressor, and the manner in which these vary from stage to stage, at one operating point only. With changes in the non-dimensional rotational speed, the pressure ratio and the mass flow alter and consequently the operating point moves along the running line shown on the axial compressor characteristics in Figure 18. In the case of a gas turbine designed for operation at constant speed and power for most of its life, the operating point changes materially only during running up and shutting down. In other applications, however, non-dimensional speed changes are frequent since this is linked to power output variation. Consequently the temperature

and pressure levels are likely to be continually varying in the compressor. In an aircraft gas turbine there will also be the changes in entry pressure and temperature produced by altitude variations which will also affect the pressure and temperature levels throughout the compressor.

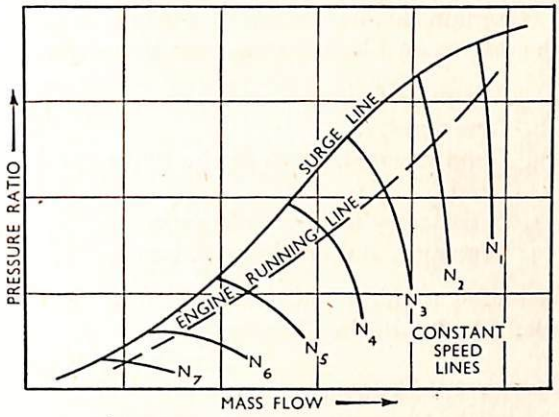


Fig. 18.—Axial compressor characteristics

When there is changing mass flow, pressure and temperature, there is likely to be turbulence in the compressor. These aerodynamic disturbances give rise to periodic forces which act upon the compressor blades and cause vibration. Since for efficient operation it is desirable to keep the ratio of blade thickness to chord comparatively low, the blade does not possess much inherent damping capacity. If the frequency of the vibrating forces approaches the natural frequency of vibration of the blades, quite large amplitudes may result. This tendency to blade vibration makes a material of high damping capacity attractive for blade construction since it will help to minimise blade failures from fatigue. Vibration in itself, of course, will not imply a likelihood of fatigue failure. It is the magnitude of the stress set up by the combination of the frequency and amplitude of the vibration which is significant.

Furthermore, since the clearance between compressor blade tips and the casing has some effect on performance and is usually small, the creep strength of the blade material must be sufficient to prevent any significant alteration of the tip clearance during the life of the compressor.

Since there is generally some likelihood of the ingestion of foreign bodies at the compressor intake, some degree of ductility is necessary in the blade material, since total blade failure could be the cause of much greater secondary damage. Small particles which may not

have sufficient momentum to cause impact failure may nevertheless give trouble from erosion if the number of such particles is high.

At the air temperature existing in the primary stages of the compressor there is also the possibility of water vapour giving rise to corrosion or pitting of the blades. Some degree of corrosion resistance in the blade material is therefore a necessity.

The main factors, therefore, governing the material selection for the compressor blading may then be summarised:

1. Adequate fatigue strength.
2. Low density.
3. Good creep strength at the temperatures likely to be encountered.
4. Satisfactory impact resistance.
5. Corrosion and erosion resistance.

How these requirements are fulfilled by the various materials available is dealt with in Chapter Five.

Combustion Chambers

Combustion intensity is very high in the gas turbine and heat releases of the order of $2{\cdot}5 \times 10^6$ C.h.u./ft^3/s. may be attained. With such rapid heat releases it becomes a problem to ensure that all the gas attains a uniform temperature. Turbulence is essential to achieve rapid mixing and therefore a uniform gas temperature is difficult to attain without some degree of disturbed flow. Unfortunately turbulence is accompanied by a pressure loss, and this has the effect of lowering the efficiency and power output from an engine so that some compromise has to be reached. Since there is a limit on top temperature due to material strength considerations it is necessary considerably to reduce the gas temperature from the level existing in the flame. It has been found that combustion is more efficient at air/fuel ratios in the 20:1 region, which is only slightly in excess of stoichiometric requirements. The flame temperature reached by combustion of fuel and the chemically necessary amount of air is in the region 2000°C (3630°F) so that for efficient burning it is necessary to have a high-temperature flame in which the combustion of the fuel is completed and then subsequently to reduce the gas temperature by the admixture of sufficient quantities of cool air.

The flame contains carbon particles which are incandescent so that the emission of radiant heat to the walls of the flame tube would raise the wall temperature above safe levels were they not continuously cooled by the flow of secondary air which is then used to dilute the combustion products. This cooling air is coming from the

compressor and is therefore at compressor delivery temperature so that the degree of wall cooling achieved depends on the temperature of the air leaving the compressor. This film of secondary air also reduces transmission of heat from the hot gases to the walls by convection, but has little effect on reducing the radiant heat input from the flame. By abstracting heat through conduction it does, however, keep the walls fairly cool. Where a steady state is attained the operating conditions are somewhat less arduous. In the industrial gas turbine this is substantially the case, but in the aircraft gas turbine there is likely to be a continually changing wall temperature. In the latter, as the forward speed and altitude change so does the pressure and temperature of the air entering the combustion chamber. This results in variations in the aerodynamic conditions in the combustion chamber. Even if the flame temperature remains sensibly constant the flame tube wall temperature is affected by changes in the temperature of the cooling air, while always there is also the possibility of a flame blow-out at altitude producing conditions of thermal shock.

In all types of gas turbine, even where a top temperature control prevents the gas temperature from exceeding a fixed level there is likely to be some degree of overshoot and in conditions of varying power output temperature cycling occurs. A change in configuration of the flame will alter radiant heat transfer conditions to the flame tube and hot spots may occur. This produces local distortion in the material and may lead to cracking and buckling of the material. "Kinematic" construction[1] can do much to take up the thermal expansion but the local variations in temperature cannot always be prevented from causing damage.

Apart from those stresses which result from differential thermal expansion, the flame tubes may be considered as relatively free from high stress conditions. In view of the very high temperatures which may be set up, however, a most important material requirement is oxidation resistance. A good thermal conductivity value in addition would help to minimise the formation of local hot spots.

The Turbine

Since it is the turbine entry temperature which determines the performance level in the gas turbine, the temperature environment in this portion is thus the most arduous from the material aspect. There is a constant trend for temperature levels to be pushed higher

[1] Kinematic construction has been defined as the maintenance of the relative positions of two fixed points in adjacent components subject to differential thermal expansion, without resisting the expansions. As the name implies, free expansion or movement of the components is allowed to occur without any change in their positions relative to some fixed plane or planes.

and higher in an attempt to improve output. Blades, discs and casings all operate at or near the maximum temperature conditions. The stator blades and the casing are not exposed to the influence of centrifugal stress as are the discs and rotor blades, and it is the occurrence of these stresses which gives rise to the most difficult material problems. For some years the difficulty of matching material to operating conditions produced a crop of failures both of blades and discs. As knowledge of operating conditions increased, however, and material development advanced the failures experienced became rare and satisfactory life is now obtainable in these components.

Although knowledge of working conditions of steam turbine discs is quite extensive it was not directly applicable in the gas turbine even though the temperature levels are not so very different. In the steam turbine a more or less steady state of temperature distribution is reached after some period of running and investigations have shown that the disc takes up a temperature approximating to that of the working fluid. In the gas turbine, however, not only are rotational speeds much higher, particularly in the case of the aircraft gas turbine, but the temperature distribution in the disc is likely to vary considerably. In most types of gas turbine cooling of one or both disc faces is common, usually with air tapped from the compressor and delivered along the bore of the rotor. Since there is heat input at the disc rim the continual flow of heat from rim to centre takes place and the temperature difference may be as high as 200°C (392°F). The temperature gradient moreover is not likely to remain the same since changes in load are met by changes in the gas temperature, so that not only does the heat input at the rim of the disc vary, but as the temperature of the cooling air also varies slightly with engine speed, the temperature distribution through the disc will be continually varying. One consequence of this marked temperature gradient is that it produces a variation in the physical constants of the material, i.e. Young's Modulus and Poisson's Ratio, and in this sense the material is not isomorphous. Thermal stresses are set up in the material which tries to expand non-uniformly under the influence of the differing temperatures. If the stresses set up are sufficiently high, plastic flow may result until stress relief has taken place.

These temperature stresses are accompanied by centrifugal stresses set up by rotation of the disc. The cross-sectional design of the disc determines the distribution of these stresses since the centrifugal loading varies as the square of the radius. A design is often chosen so as to give approximately constant centrifugal stress over the whole cross-section, as shown in Figure 19.

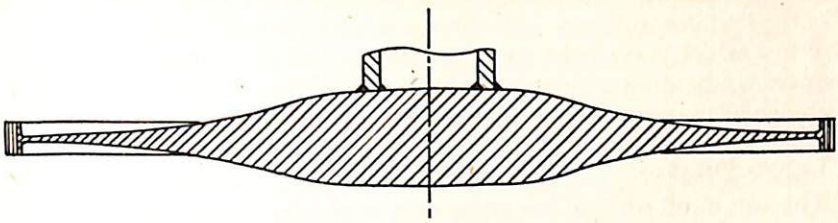


Fig. 19.—Cross-section of a disc designed for approximately constant centrifugal stress over the section

The magnitude of the centrifugal stresses will be governed by three factors, namely the speed of rotation, the dimensions of the disc and the material density. Furthermore bending stresses are imposed on the disc due to the transmission of thrust from the blades, and since this type of stress is accentuated each time the blade passes a stator blade passage there is a fluctuating stress superimposed on the steady centrifugal stress.

From this brief survey it can be seen that the materials used in the turbine must have the maximum possible resistance to creep and fatigue stresses at temperatures in the 600–800°C (1112–1470°F) region, coupled with adequate oxidation resistance at this temperature level. Other desirable features include good thermal and mechanical shock resistance.

In this section, the material requirements of the designer have been briefly surveyed. In later chapters the further requirements imposed by the component manufacturer will be discussed.

(Section Two)

The Influence of the Fuel on Operating Conditions in the Gas Turbine

An inherent advantage of the gas turbine is its potential ability to burn a very wide range of fuels. In the absence of a very large increase in the operating temperature, it seems unlikely that the overall efficiency of this new prime mover will be significantly greater than engines of the Diesel type or the steam turbine. Consequently the future of the gas turbine may be largely dependent on its utilisation of the cheaper kinds of fuel.

Solid, liquid and gaseous fuels have all been utilised to some extent in gas turbines with varying degrees of success where there

was an economic advantage in departing from the kerosene used in the first gas turbines. In this section a survey of the various types of fuel which have been used, or may be used in the near future, is made, indicating any special influence of particular fuels on the choice of materials of construction.

Liquid Fuels

The use of oil and oil products as a source of power has increased enormously in the present century and the consumption of this natural product now ranks second only to coal. From time to time it has been suggested that the tremendous amount of petroleum being used each year will rapidly exhaust all the world deposits but the introduction of scientific methods of prospecting has led to the discovery of many new deposits and at present, at any rate, the supply seems to be assured for many years to come.

Mineral oil is, like coal, found in sedimentary rocks. It is believed to originate mainly from the action of bacteria on organic material in stagnant waters, where the atmospheric oxygen did not penetrate. The bacterial activity resulted in the formation of the heavy hydrocarbons and the subsequent formation of the lighter paraffins probably occurred as a result of the pressure and temperatures arising during the burial of the material by the accumulation of sediments and strata movements. This crude oil gradually drained from the strata where it was formed and tended to collect in particular locations where the over- and under-lying rocks formed a natural and impervious trap. The gaseous decomposition products were sometimes also prevented from leaking away and remained in the space above the oil pool. The use of this "natural gas" will be discussed later. There may also be trace quantities of a wide variety of metals, which exist as mineral salts derived from the strata containing the petroleum or as the oil soluble salts of petroleum acids. The commoner of these metals include vanadium, nickel and iron. Water is often present in an emulsified state and in amounts as high as 30% of the crude. Most crudes also contain sulphur in amounts ranging on an average from 0·5% to 3%, but some may have a considerably higher proportion, as in the case of an Iraq crude which contains over 7% sulphur. Conversely there are low sulphur crudes having less than 0·5% but these are becoming rare.

The crude natural oil is a mixture of solids, liquids and gaseous substances. In general the deposits can be designated either as paraffinous or naphthenic, although mixed deposits are also common. A paraffinous petroleum contains large quantities of paraffin wax while the naphthenic types have little paraffin wax but a lot

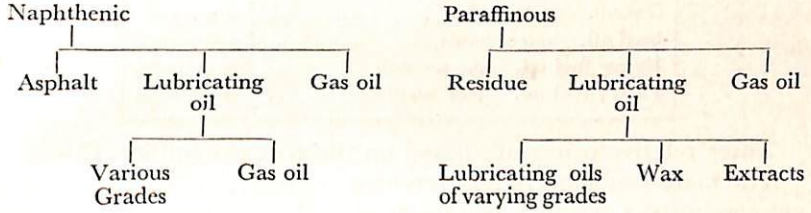
of asphaltic matter. The gaseous portion, which contains the liquefiable gases, e.g. butane, is a valuable by-product of many of the wells and is discussed in the later section on gaseous fuels.

The crude petroleum undergoes special refinery processes and fractional distillation, resulting in the production of several different materials.

The initial distillation separates the crude petroleum into:

(a) Topped crude.
(b) Gas oil.
(c) Kerosene, otherwise known as paraffin oil.
(d) White spirits, solvents, etc.
(e) Petrol, or gasolene, of varying grades of volatility.

The "topped crude" fraction is further separated according to the nature of the oil, viz.:



The petrol and gas oil fractions provide the main fuels for internal combustion and compression-ignition engines. Certain of the distillate and residual oils are used as fuel oils, either separately or after suitable blending, though the more viscous oils require some degree of pre-heating for use. Of this wide range of oil products the internal-combustion engine operates mainly on those in Group (e) and the compression-ignition engine on those of Group (b). The steam turbine may be operated by the burning of the heavy fuel oils in the steam generating plant. The gas turbine, however, is capable of producing energy directly from the combustion of a number of these products ranging from kerosene to the low-grade "residual" oils, i.e. a fuel containing a proportion of refinery residues. Petrol is, of course, too expensive and too explosive to form a satisfactory fuel.

Ever since the early days of the gas turbine kerosene has been used in aircraft power units. The reasons for this choice are mainly

(a) Kerosene is widely available at a reasonable price.
(b) The danger of fire is appreciably lower than that of the other aviation fuels.

(c) The fluidity, even at the ice-point, makes any pre-heating unnecessary.

(d) Kerosene is satisfactory for the ignition requirements and burns with a steady flame.

In the case of land-based or marine gas turbines kerosene is too expensive for all but the low-power units. At the present time various grades of distillate oil are being used in the majority of the higher power units but the developments now taking place in the use of the heavier oils will eventually lead to the adoption of such fuels. This could result in an appreciable lowering of the running costs of a gas turbine, as can be seen from the following table:

TABLE V

Relative prices of potential gas turbine fuels in the United Kingdom in 1955

Aviation turbine fuel	1·0 –1·6
Gas oil. 35 seconds	2·9 –3·4
Fuel oil. 200 seconds	0·65–0·71
Heavy fuel oil. 950 seconds	0·51–0·63
Heavy fuel oil. 3500 seconds.. ..	0·48–0·58

These relative prices are based on the cost per million C.h.u.

The main problems to be overcome are concerned with the efficient combustion of such a low-grade material and the corrosive effect of the combustion products, particularly when the sulphur and vanadium contents of the fuel are high. This is discussed in more detail in another chapter.

Availability is a further important factor in the selection of a fuel for marine use. Since ships cannot be expected to rely entirely on their home port supplies the fuel must be readily available at the majority of the ports. The cost of the fuel must also be as low as possible. If the marine gas turbine is running on Diesel oil the cost of the fuel is higher at the present time than in the equivalent Diesel engine, since the volumetric fuel consumption for similar output is greater in the former. Consequently, to introduce the gas turbine as an economical power source, the use of a cheaper fuel than Diesel oil is essential. The residual oils are in this category but again problems of efficient combustion and corrosive effects are encountered.

The ash-forming constituents of a crude oil may be of the order of several per cent and these become concentrated in the residues during the distillation process. Consequently, although the wholly distillate fuels have a negligible ash value, generally less than 0·01%, any residual fuel has an appreciable ash content, generally between

0·05% and 0·1%. The importance of this value will be appreciated when it is realised that, using a fuel producing 0·1% ash at a rate of 1000 lb/h, the rate at which ash passes through the turbine is approximately one ton every two thousand hours.

The substances present in the ash are mainly the oxides or sulphates of metals such as sodium, calcium, aluminium, magnesium, vanadium, iron, copper, nickel, etc. Certain crude oils, notably those from South America form ashes containing a very high proportion of vanadium oxide and the serious blade corrosion which can be experienced in the gas turbine when vanadium-containing oils are used is discussed in Chapter Three.

When the danger of severe corrosion arising from the vanadium was recognised, intensive research was initiated by the oil companies in order to discover in what form the vanadium is present in the crude oil, and whether any cheaper method than distillation can be found to effect its removal. The results so far reported are not encouraging. The vanadium is present as complex organic compounds which are readily soluble in the crude. Treatments such as centrifuging and filtration are consequently ineffective. Sodium and calcium have however been successfully reduced from 80–120 p.p.m. and 150–250 p.p.m. respectively to about 4–7 p.p.m. in each case by a combined washing and centrifuging process which is claimed to be sufficiently economical to justify itself (1).

On the other hand, the suggestion has been made that by permitting a slight loss in combustion efficiency of the order of 1%, the ash particles will remain enclosed in a matrix of unburnt carbon and will thus be rendered innocuous. Varying degrees of success have been achieved in this way but it seems likely that the increased risk of fires in the heat exchangers, caused by the presence of these carbon particles, makes this method rather impracticable.

One other trouble which arises in the use of the heavy fuel oils is the deposition of carbon on the fuel injection tubes and the walls of the flame tube. The origins of this carbon are twofold; quenching of burning particles in the later stages of combustion and the "cracking" of unburnt fuel on the very hot metal parts. The latter will obviously be more prevalent at the fuel entry, and will be accentuated by poor atomisation, over-rich mixtures and the use of fuels of high aromatic content. The presence of this carbon can lead to carburisation effects in the underlying metals and serious loss of mechanical properties can be caused.

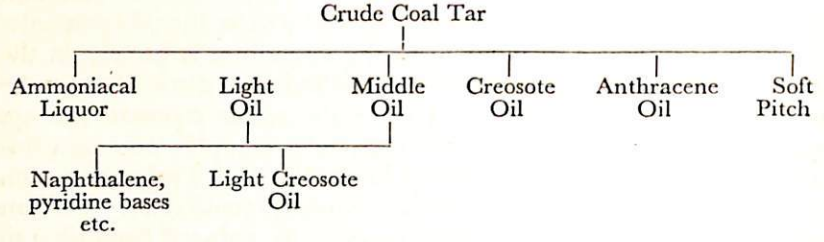
Coal Tar

This material is a very useful fuel for the gas turbine since it contains no very serious impurities. It is unlikely to become an important

gas turbine fuel since it is only available in limited quantities and there are other uses to which it may be put. There are, however, several turbines now in operation using this fuel, including one at the Bank of England, Debden.

Coal tar is mainly a by-product of the high-temperature carbonisation of coal, although a small amount is obtained by low-temperature carbonisation. Although the yield of coal tar varies considerably with the nature of the coal and the carbonisation conditions, about 10–20 gal/ton of coal is an average figure.

The crude coal tar is subjected to a distillation process which yields a number of products:



A number of fuels have been obtained by blending these various products in different amounts and for convenience these fuels are designated by a number which is the temperature in degrees Fahrenheit to which the fuel must be pre-heated for satisfactory atomisation and combustion. Thus the soft pitch residue is referred to as C.T.F. 400, i.e. a coal tar fuel which must be pre-heated to 400°F. By blending the creosote oils and anthracene oils, with or without the soft pitch, five main grades of fuel are produced and these are known as C.T.F. 50, 100, 200, 250 and 300.

The sulphur content of these fuels does not exceed 1% while the ash content ranges from less than 0·05% in C.T.F. 50 to 0·75% max. in C.T.F. 400. The ash is mainly ferric oxide together with lime and silica. Unlike the oils derived from petroleum, no vanadium is present. The calorific value of these fuels ranges from 16,000 to 17,000 B.t.u./lb.

For efficient combustion good atomisation is essential. The flame temperature is in the region of 1700°C (3092°F) and the flame is very luminous, so that a high rate of heat transfer is experienced. In the gas turbines constructed to use this fuel, it has been found preferable to use a refractory lined combustion chamber so as to avoid the problems which would arise with such a luminous flame in the cooling of metal flame tubes. The use of refractories also aids efficient combustion since a greater concentration of heat can be permitted. Since the ash is high in ferric oxide, the use of silicious

refractories has to be avoided to prevent the formation of low-melting-point iron silicates.

Gaseous Fuels

The majority of oil-fields are rich in sources of "natural gas", which collects in the rocks above the oil pools. In addition to the mid-Continental and Californian oil-fields of America, the gas is extracted on a commercial scale in the oil-fields of Italy, Holland, Germany, France, Hungary and Austria. The composition of this gas varies over wide limits, but the percentage of saturated hydrocarbons, especially methane, is usually high. A typical average analysis of the gas is:

80% methane, CH_3, 6% ethane, C_2H_6, 3% propane, C_3H_8, 1% butane, C_4H_{10} and 10% nitrogen.

Since the gas contains such a small percentage of non-combustibles, the calorific value is high, about 700–900 B.t.u./ft^3, and it is not unusual to find the gas being piped over several hundreds of miles to industrial centres.

The power needed for the pumping of the gas over these long distances can conveniently be supplied by the use of a medium-power gas turbine, since this engine is readily operated on this type of fuel. The small amount of fuel consumed is negligible when compared to the large quantities passed by the pumping units. Already a large number of these gas turbines have been introduced on the American pipe lines and are reported to be operating satisfactorily and economically. With this type of fuel few impurities are present and consequently problems in the choice of materials of construction for the gas turbine are not made more difficult from this aspect.

A similar type of fuel is obtainable from some coal seams, where the nature of the strata is such that large quantities of methane have been entrapped during the organic decomposition. The mining of such seams involves the problem of removing this dangerous gas. If the leakage of gas from the seam at the coal face is slow, it is usually possible to deal with the removal problem by installing powerful fans to supply air to the workings. The resulting air and gas mixture blown from the pit is extremely dilute, containing only about 1% methane, but the very low calorific value of only 5–10 B.t.u./ft^3 is offset by the very large quantities, of the order of several million cubic feet, produced daily at each mine. Although 1% represents a suitable final air-to-fuel ratio for the gas turbine, it is too weak for initiating combustion and experimental turbines are in operation at the present time using this fuel, suitably enriched

by other fuels such as town's gas, oil, coke-oven gas or producer gas. Ultimately it is hoped to utilise the large quantities of low-grade "fines" produced in the coal cleaning plant as auxiliary fuel.

In seams where a high gas evolution rate is encountered, however, the forced draft method of ventilation is being superseded by a process known as "methane drainage". Small bore holes are drilled into the coal seam some distance ahead of the main mining operations and pumping engines draw off the gas in large quantities. The concentration of the methane in the exhaust from these pumps may be as high as 30% and the gas forms a valuable source of electrical power for the pit-head machinery when used in a gas turbine driving a generator.

Since methane is the main gas produced during the decomposition of organic material, it is not surprising to find that methane-air mixtures are also available, and used for power production, in the large sewage works. There are some recent instances of specially designed gas turbines being installed for power production from these "sludge gases" as they are called, and it is certain that future installations will find this new prime mover more suitable than those now used.

A scheme which is the subject of a great deal of development work at the present time concerns the gasification of coal seams *in situ*. The generation of a gaseous fuel from coal without the necessity of mining has exercised people's minds for many years, experiments being recorded in Russia as early as 1932. Other countries followed in more recent years and pilot plants are now in operation

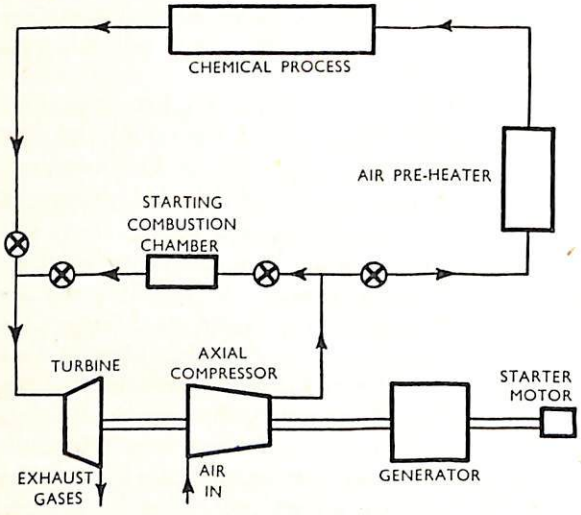


Fig. 20.—A simplified flow diagram for the Houdry catalytic cracking process

in this country at Newman Spinney, near Chesterfield and Bayton in Worcestershire.

The underground gasification process was at first visualised as replacing the normal mining operations, but it is apparent from the pilot operations that the process is only suitable for the thin, wet or dirty seams which cannot be economically used in any other way. Such seams are widespread in the British Isles and it has been estimated that the development of a full-scale process would make use of more than 1000 million tons of coal. Unfortunately there seems to be little prospect of recovering a gas with a high heat content, so that for economic reasons the gas must be used very close to the source of production.

Briefly the operation of underground gasification consists of drilling two channels down to the coal seam some 50 ft apart and igniting the coal by one of several available methods. By pumping air down one tube the coal may be burnt under reasonable control so that the gas issuing from the second pipe consists of the products of several reactions, e.g.:

Producer gas reaction: $2C + O_2\ (+4N_2) \rightarrow 2CO\ (+4N_2)$
Water gas reaction: $C + H_2O \rightarrow CO + H_2$

The water for the second reaction is nearly always present in the coal itself, but steam could be supplied if necessary. In addition, the hot coal will give off coal-gas (mainly methane, hydrogen and hydrocarbons) and unless these burn in the air stream a certain amount will be present in the issuing gas.

The main difficulty associated with underground gasification by this method lies in the fact that, as the channel through the seam widens by the combustion of the coal, the air stream tends to pass through without reacting with the coal and a very weak gas results. The calorific value of the gas may consequently vary from the initial value of about 100 B.t.u./ft^3, to as low as 10 B.t.u./ft^3, in the later stages. The highest value is less than one-quarter of that of town's gas, mainly due to the vast quantities of inert nitrogen included. The enrichment of the air with oxygen or even the use of pure oxygen itself is likely to prove economically impossible in full-scale production although by such means the heat content of the gas could be raised by a considerable amount, probably to 250–300 B.t.u./ft^3.

The uses for a gas of this type are either the synthesis of petrol, probably using the Fischer-Tropsch process, or the generation of electricity. It has already been announced that electricity has been generated on the Bayton site using a reciprocating gas engine. This type of engine is unsuitable, however, once the calorific value of the

gas falls below some 70 B.t.u./ft^3, since at this level normal ignition is impossible. The gas turbine is then the only engine capable of utilising such a low-grade gas and this is being actively pursued.

In addition to these natural fuels, there are occasional waste products in industrial processes which may be used as a fuel for the gas turbine.

One example of this type of application is in the Houdry catalytic process, used for the production of high-octane gasolene. In this process large quantities of pre-heated air at high pressure are used in the cracking of crude petroleum and in the early days of the gas turbine it was realised that by the use of this new engine an appreciable economy could be affected. Figure 20, on page 50, shows a simplified flow sheet for this process. The air is compressed to some 50 lb/in^2 and pre-heated to the temperature required by the catalytic process. The exit gases, at some 510–540°C (950–1000°F) and 45–50 lb/in^2, pass through the gas turbine. This abstracts sufficient energy from the gases to drive the compressor and also an electrical generator, releasing the gas at about 300°C (527°F) to be used in a steam generator.

A further example is of a closed-cycle gas turbine making use of the heat in the waste gases from vertical retorts, which is installed at the Foleshill works of the Gas Board. A waste heat boiler generating process steam is used to extract the residual heat in the waste gases issuing from the gas turbine heat exchanger.

Blast-furnace gas is a valuable by-product of iron-making, and, since its use in the furnace air pre-heaters is limited to some 20–25% of the amount produced, a considerable proportion is available for power production. The gases issuing from the top of the blast-furnace contains a fairly high proportion of carbon monoxide together with a little hydrogen, along with non-inflammable gases such as carbon dioxide and nitrogen. The analysis of the gas varies widely according to the conditions in the furnace.

The heat content of the cooled and cleaned gas averages approximately 100 B.t.u./ft^3 (1200 B.t.u./lb), and a furnace will produce a quantity of gas equal to some five or six times the weight of iron, so that about 12 million B.t.u. are available for each ton of pig iron produced. The cleaning equipment of a modern works reduces the dust content of the gas to less than 0·01 grains/ft^3 and it may be as low as 0·005 grains/ft^3. At this level it presents no problems in the gas turbine, and may be used at an air-to-fuel ratio of about eight or nine to one.

At the present time the majority of the surplus gas is used either in waste-heat boilers or in gas engines, but already there are one or two instances on the Continent of this gas being used in a gas turbine,

Plate 6. *A flame tube from the early Bristol Proteus* 600 *series turbo-prop, illustrating buckling that has occurred during development running*

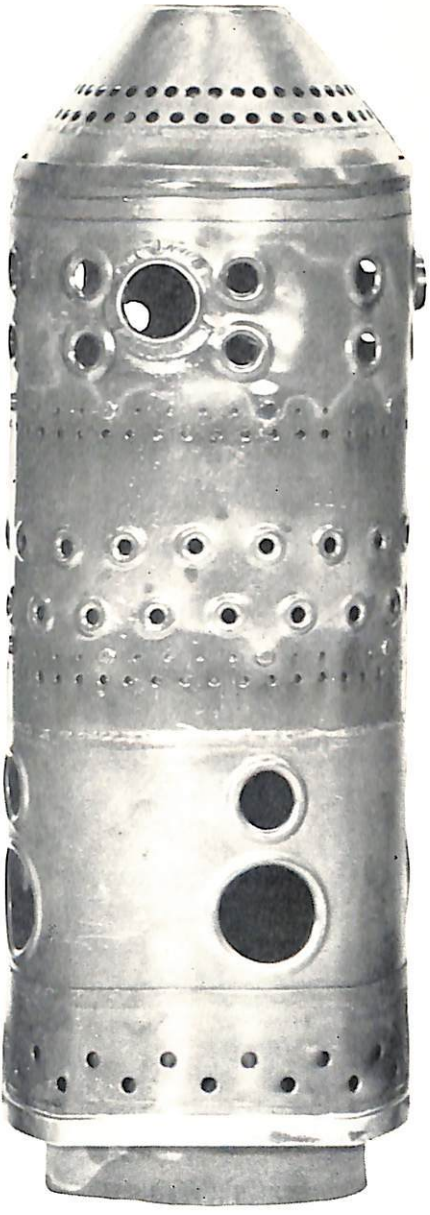


Plate 7. *A flame tube from the later Proteus* 700 *series engine, indicating the good condition of the component after a* 150*-hour type test run*

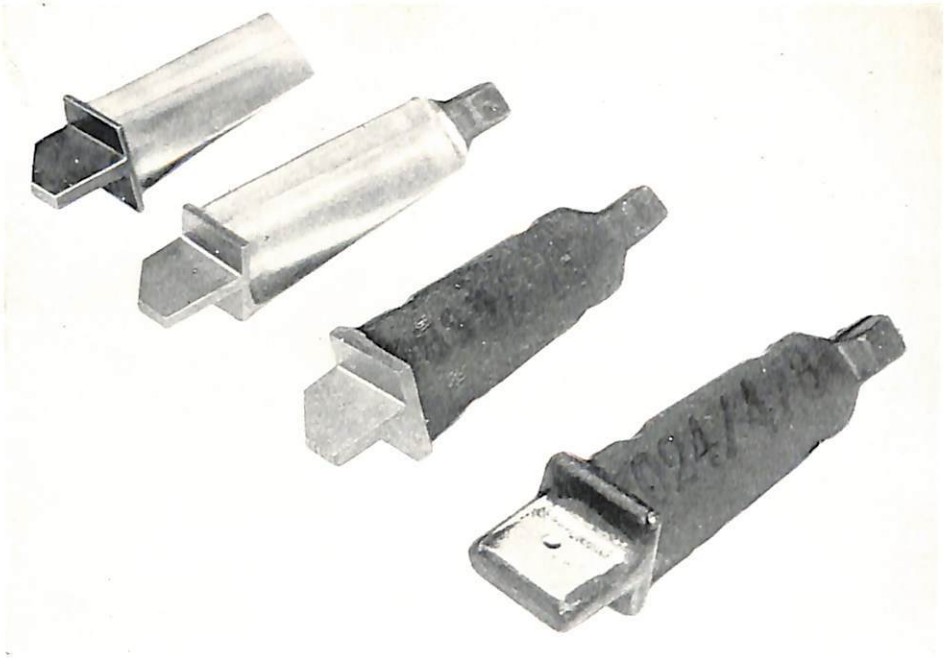


Plate 8. Stages during manufacture from a forging of Armstrong Siddeley Sapphire blades

Plate 9. Stages in production of a blade forging from bar stock

for example at Dudelange Steelworks, Luxembourg and Barecaldo Steelworks, Spain. One gas turbine is also in course of construction in this country to operate on blast-furnace gas.

Solid Fuels

The adaptability of the gas turbine has led to many experimental units being designed for solid fuels and the knowledge thus obtained will undoubtedly greatly speed the introduction of a number of commercial models.

At the present time two solid fuels are receiving considerable attention, namely coal and peat. Both these materials are freely available and used in the generation of power by steam turbines. The advantage of using them in the gas turbine would lie in the abolition of the expensive steam generating units. Peat, especially, is usually unsuitable for commercial application, but is used for steam-raising at power stations in Ireland where, since coal would have to be imported, there is a considerable economic advantage in the use of this low-grade material.

The use of a solid fuel presents certain special problems mainly connected with the provision of efficient combustion conditions and ash removal equipment. It was soon found that if the ash particles were not removed before the hot gases entered the turbine, rapid erosion of the blades occurred and the deposition of the ash in the passages caused serious fouling.

The closed-cycle gas turbine, would thus appear to offer several advantages, especially the complete isolation of the moving parts from the ash and corrosive gases. This is offset to some extent, however, by the problems encountered in the construction of a heat exchanger to operate under these conditions. The deposition of a layer of fine ash on the tubing can cause an appreciable loss of efficiency in heat transfer unless periodically removed and may give rise to corrosion of the metal. In addition, the fouling of the gas passages between the tubes sets up a back pressure which would eventually give rise to surging or stalling in the compressor.

A second method aimed at overcoming the problem of ash removal, the exhaust-heated cycle, is being tested for coal-burning gas turbines. In this case the products of combustion again do not enter the turbine but pass through a heat-exchanger. There they give up their heat to the secondary flow which consists only of air, and it is this hot air which is then expanded in the turbine. By the use of this cycle the problem of erosion and ash deposition in the turbine is replaced by that of maintaining a clean heat-exchanger.

Coal

As stated above the removal of ash from the hot gases presents the greatest problem to be overcome in the use of coal. At first sight, it would seem that this problem could be avoided by first gasifying the coal and then using the coal gas in a conventional combustion chamber. The size and initial cost of the gasification unit and accompanying gas-cleaning equipment is, however, generally prohibitive. Consequently a great deal of attention is being paid to the question of ash removal from the products of combustion of a stream of pulverised or granulated coal.

The pulverising of coal dates from the latter years of the nineteenth century and was originally pioneered to make use of the dirty low-grade fuels. It is now recognised that in many applications the pulverised solid fuels can be burnt more efficiently than large pieces, and it is common practice to pulverise even high-grade coal for industrial use.

The ash content of the coal to be pulverised is of importance in assessing the degree of fineness required, since where ash and coal are present in the same particle, combustion will be greatly prolonged. In addition, although the smaller the particle the more efficient the subsequent combustion will be, the cost of the additional power required for the milling process sets a practical limit to the size. It is probable that if 90% of the particles pass through a 200-mesh sieve, the best compromise is reached between the various factors.

The actual burning of the pulverised fuel in a combustion chamber presents many problems. Since the combustion is a surface reaction, the speed of burning is controlled by the rate at which the products of combustion are removed from the surface and replaced by a fresh air supply. A pound of coal requires approximately 10 lb of air for total combustion, or over 10,000 times its own volume, from which it can be seen that the ample provision of air to the burning particles is an important feature of rapid combustion.

Since such small particles tend to move at the same speed as any air stream to which they are introduced, this problem of air supply is rendered more difficult since, under such conditions, the only way new air can reach the surface is by diffusion through the layer of combustion products. Consequently several ingenious methods have been devised for introducing highly turbulent conditions.

After combustion has finished, the solid particles of ash have to be removed before they can damage the turbine blading. In the development of the Allis-Chalmers' locomotive gas turbine it has been reported that ash particles with a diameter greater than approximately 20 microns caused very rapid erosion in the primary

blading stages, and it is probably desirable to free the gases from particles of over 10 microns diameter to ensure freedom from this trouble.

In the present experimental coal-burning plants both metal and refractory-lined combustion chambers are being used. The metal type have the advantage that local overheating and ash adherence are minimised, though even so a certain amount of distortion seems to have been experienced. The chamber is usually constructed from sheet metal, by welding a cylinder to a conical section, the fuel injection equipment being located in the apex of the cone. An oil pilot flame is required to maintain ignition of the solid fuel since, with the present types of burner, trouble is experienced with the flame blowing out at high loads and flashing back at low loads.

The use of refractory brickwork as a lining to the combustion chamber is advantageous in that, by preventing heat losses and maintaining a higher temperature level, the efficiency of the combustion process is increased; but the linings possess the usual disadvantages associated with refractory materials, such as susceptibility to thermal or mechanical shock and slagging.

Two main designs of combustion chamber have been developed called the vortex and the cyclone types:

Vortex Combustion Chamber

The first chamber of this type was designed by T. F. Hurley, of the D.S.I.R. Fuel Research Station about 1930, following observations by Dr. Lander, then Director of Fuel Research, on the movement of solid particles in a free vortex. Certain particles, instead of travelling to the centre as may have been expected, rotated in orbits at a constant distance from the centre, presumably due to balancing of the inward drag of the vortex by the outward forces due to the centrifugal motion. It was suggested that if ignited solid fuel particles were injected into an air vortex they would remain in the chamber at a position governed by their size, being constantly "scrubbed" by the air stream until combustion was complete. Although the potentialities of the design were appreciated, it was neglected at that time for use in boilers in favour of other designs. On the introduction of the solid-fuel gas turbine, however, the idea was revived. The initial combustion of the fuel is generally begun in a separate chamber and the burning particles then injected into the periphery of a swirling air stream. On detailed study, the position taken up by a particle was found to be governed by a number of factors and the balancing of the various forces is certainly more complex than was at first thought. The hot gases leave the chamber from the centre and pass into a dust separator.

The choice of a refractory lining in this type of chamber is restricted by the slagging action of the hot ash particles, while metal-walled chambers tend to be distorted by the uneven temperature distribution in the chambers so far tested.

Cyclone Combustion Chamber

By contrast with the vortex design the cyclone type concentrates on causing the solid fuel particles to adhere to the walls of the chamber so that the air stream will sweep over them. The particles then burn there and the ash forms a fluid slag with the refractory lining. This design has been used for many years in conjunction with water-tube boilers but many modifications are necessary for its application as a gas turbine auxiliary. The main development work in this country is being carried out by the British Coal Utilisation Research Association, and two experimental cyclone combustion chambers have been operated with promising results.

The chambers, approximately $2\frac{1}{2}$ ft diameter $\times$ 10 ft high, are operated in a vertical position instead of the inclined position preferred for boiler work, and the air supply enters from the top and the bottom respectively in the two units. The air stream enters tangentially to the periphery of the cylinder and swirls round the walls before leaving at the opposite end. Granular fuel, passing a $\frac{1}{16}$-in. mesh, is introduced into the air stream and is thrown into contact with the slag-covered walls. The particles stick and burn rapidly in the constantly changing air conditions. The refractory wall is water-cooled in the present units, but since this represents a considerable heat-wastage, this would be altered to serve as an air pre-heating unit in actual practice. The walls waste away under the slagging conditions to an equilibrium condition in which the ash is deposited at a rate equivalent to the rate at which fluid slag drains away down to the bottom of the chamber. The issuing gases are cooled by the admission of more air before entering a dust separator. At the present time combustion efficiencies of over 95% have been achieved with approximately 90% ash removal.

Both the above units operate at atmospheric pressure which is unsuitable for use in conjunction with a gas turbine, but the development of a slagging cyclone combustion chamber to operate with compressed air is in progress.

Research into the problems involved in burning solid fuel is also going on in Australia. Very large deposits of brown coal are to be found there and the development work has so far been mainly concerned with the combustion of this fuel.

Brown coals are roughly intermediate between peat and coal. Although they have a high moisture content, about 30–50%, the

original fibrous structure is almost completely obliterated. When dried, therefore, the material is brittle and easily pulverised. In this way it can be fluidised for feeding into the combustion chamber of the gas turbine. As with ordinary coal, the efficient combustion of the solid particles and removal of the ash particles after combustion are presenting difficult problems.

Peat

It is estimated that some 600 million tons of peat are available in Scotland alone, and consequently it is not surprising that uses for this vast quantity of fuel are being sought. The present experimental work on the use of peat in a gas turbine is highly promising in this respect.

The project is the result of co-operation among several organisations, the Ministry of Fuel and Power, the North of Scotland Hydro-Electricity Board, the Peat Division of the Department of Agriculture for Scotland, John Brown and Co. Ltd. and Ruston and Hornsby of Lincoln. In 1953 work began on a full-scale experimental plant at Altnabreac, in Caithness. The station comprises a peat de-watering plant and combustion equipment together with a 2000-kW closed-cycle gas turbine built by John Brown's and a 750-kW open-cycle turbine developed by Ruston and Hornsby.

The main difficulty to be overcome in the use of peat is to remove the large amount of water present. Peat is formed from dead plants by the action of bacteria and fungi in waterlogged environments. The latter conditions arise where an impervious bedrock is present in a region of high rain-fall, and consequently peat bogs are widespread in the northern hemisphere. Whereas in the presence of sunlight and oxygen, cellulose material is converted to carbon dioxide and water:

$$C_6H_{10}O_5 + 6O_2 \rightarrow 6CO_2 + 5H_2O$$

in the swamplike conditions the complete conversion is inhibited, the resulting product being a brown humus:

$$\text{(approx.)}\ 2C_6H_{10}O_5 \rightarrow \underset{\text{(humus)}}{C_8H_{10}O_5} + 2CO_2 + 2CH_4 + H_2O$$

In the drier seasons new plants grow on the surface of the bog and die away in their turn, so that in time many feet of the humified residues accumulate. These comprise the present peat bogs, and it is generally recognised that any project which can remove this material and open up the area to agricultural development should be fully explored.

The natural peat contains some nine pounds of water to each

pound of solids and in this condition the amount of heat required to drive off this water is greater than the calorific value of the resulting fuel. Consequently the greater part of the water must be removed by mechanical means, such as pressing. By this method it is possible to remove about two-thirds of the water, the majority of the remainder being colloidally bonded and therefore unremovable by this technique. The final water content of about 20% is obtained by subjecting the pressed material to the drying action of the hot waste gases issuing from the turbine. The peat is then shredded by special machinery before being fluidised for feeding into a refractory combustion chamber.

The fully-dried peat contains 65–68% volatiles, 30% fixed carbon and from 2–5% ash, and has a calorific value of 7000–9000 B.t.u./lb. The problems of ash removal will be similar to that in the case of coal and some separating apparatus has to be incorporated. In the Scottish trials a fly-ash separator is being used, apparently with reasonable success.

Special Fuels

In the so-called "atomic age" it is interesting to note that research work is well under way to provide a gas turbine operating in conjunction with a nuclear reactor. In America, particularly, it has been announced that the General Electric Company are working on the production of a power plant for aircraft use. The probable arrangement in such cases appears to be a nuclear reactor, operating with a core temperature of 700–800°C (1300–1470°F), cooled by a flow of liquid sodium or helium gas. If the former is used a heat exchanger then abstracts the heat from the sodium, the heated air being used to drive an open-cycle turbine. With helium gas, however, direct use could be made of the heated helium in a closed-cycle turbine. In either case, serious material problems arise, mainly connected with the nuclear reactor, but also in the use of liquid sodium in a heat exchanger. The latter is discussed in Chapter Four.

REFERENCES

1. D. M. Landis. *Trans. Amer. Soc. Mech. Engrs*, January, 1956.

THREE

Metals under Stress and Temperature

THE ADVENT of the gas turbine resulted in an intensive investigation of the properties of metals and alloys at elevated temperatures. Interest centres mainly around the mechanical properties and especially the ability to withstand the differing types of stress encountered in service. These stresses fall into three main categories, steady, fluctuating and shock stresses. Each of these may cause failure of the material and its ability to withstand combinations of all three has to be considered in selecting alloys for high-temperature service.

In discussing the behaviour of metals and alloys under these three types of stress, it must be pointed out that, while a good deal of the published information on the high-temperature alloys deals with the properties of the material under controlled conditions, and usually only under one type of stress, the stresses set up in service are more complex. Many other factors may considerably alter the behaviour of the alloy, for example, oxidation or decarburisation may result in surface deterioration and the production of minute cracks or the stress system may slowly alter the structure and mechanical properties of the alloy. Although, for simplicity it is preferable to discuss each type of stress as though it alone were acting, these reservations must be borne in mind.

STEADY STRESSES

When a steady load, sufficient to cause plastic deformation, is applied to a metal or alloy at room temperature, "work-hardening" will occur in most cases and the consequent increase in the yield point will enable the material to resist further deformation.

When, however, the same test is repeated at an elevated temperature it is found that more complex mechanisms are involved, work-hardening being less pronounced and deformation continuing over a period of time, depending on the load, until fracture occurs. The load required, for fracture in such cases, is appreciably less than that indicated by the normal tensile test. This introduction of a time factor has led to a detailed study of the processes involved in this "creep" of metals. A similar effect is observed with several soft metals, e.g. lead and zinc, at room temperature.

When a metal is subjected to a steady tensile stress at the test

temperature and the extension of the test piece noted over a prolonged period of time, a curve of the type shown in Figure 21 is obtained. Immediately the load is applied a certain amount of elastic and plastic deformation occurs, indicated by the line AB. This is followed by a period of decreasing strain rate, BC, which may be compared to the onset of work-hardening which would occur at lower temperatures. These early stages are usually referred to as primary creep. The material then enters a stage in which the strain increases linearly with time, CD (secondary creep), and, if the test is sufficiently prolonged, a point D will eventually be reached at which there is a relatively rapid increase in the strain rate (tertiary creep) until the test-piece fractures.

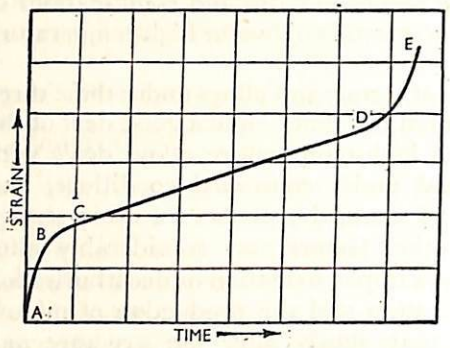


Fig. 21.—A simple creep curve in which all three stages of creep are shown

The study of the various mechanisms comprising "creep" can be traced back to the first few years of this century but the most important early work is that of E. N. da C. Andrade (1) published in 1910 and 1914. His investigations ranged over many metals and alloys, of differing crystal structures, and also a wide variation of temperatures and stresses. From his results, Andrade showed that the creep curve could be expressed by an empirical equation of the type

$$l = l_0(1 + \beta t^{1/3})e^{Kt}$$

In this equation l is the length of the specimen at time t, K and β are constants and l_0 is a further constant which approximates to the length of the specimen immediately after the initial extension has occurred.

Furthermore, Andrade showed how the creep curve could result from the co-occurrence of two different types of flow. For, from the equation, if $K = 0$, then

$$l = l_0(1 + \beta t^{1/3})$$

and

$$\frac{dl}{dt} = \tfrac{1}{3} l_0 \beta t^{-2/3} \quad \text{or} \quad \frac{dl}{dt} \propto \frac{1}{\sqrt[3]{t^2}}$$

from which it is clear that as the time of application of the stress becomes greater, the rate of extension approaches zero. This type of flow is generally referred to as β or transient flow.

On the other hand, when $\beta = 0$,

$$l = l_0 e^{Kt}$$

and

$$\frac{dl}{dt} = l_0 K e^{Kt}$$

or, since $l = l_0 e^{Kt}$,

$$\frac{dl}{dt} = Kl$$

and the rate of extension is therefore approximately constant if the slight change in the length of the specimen during creep be ignored. This type of flow has become known as K or steady-state creep.

The experiments also showed that the β term is more important at the lower stresses and temperatures while under more severe conditions the K flow predominates. The effect of temperature on the steady-state creep rate is now known to follow an equation of the type.

$$\log (\text{creep rate}) \propto \frac{1}{\text{absolute temperature}}$$

where the applied stress remains constant.

Primary Creep

The extent of the initial deformation which occurs immediately the load is applied will depend on the nature of the material under test and the chosen conditions of temperature and load. The properties of a material in this respect can be fully determined by the normal tensile test carried out at the required temperature.

The transient creep stage is generally associated with the appearance of "slip bands" in the grains of the metal. In any crystal structure there are certain atomic planes which can be shown to be slightly weaker under stress than others. Consequently if the crystal is subjected to plastic deformation it is found that these weaker planes will fail by shear. Under the microscope this is indicated by a series of parallel bands across the grain and these are referred to as "slip bands". In addition it has been shown that during this stage of creep the metal crystals tend to rotate slightly to re-orientate themselves in the direction of the applied stress.

These two processes are usually associated with the phenomenon known as "strain-hardening" which brings about an increased resistance in the metal to further deformation. The strain-hardening set up does not seem to be rapidly dispersed by thermal softening effects, as might be expected, since a large proportion of the strain observed during the period of transient creep is recoverable if the stress is removed and sufficient time allowed for the reverse process to occur.

Consequently the suggestion sometimes put forward that the occurrence of this strain-hardening effect and its counter-balancing by thermal softening may be the reason for the decreased strain rate and subsequent steady state of creep is unlikely to be correct, except in a few special cases, but these two effects will certainly participate to some extent. Modern theories concerning the presence and properties of dislocations, i.e. faults, in the structure of a metallic crystal have been developed to a stage where it is possible to predict their probable behaviour when stresses are applied to the crystal and from this, by making various assumptions, to derive a mathematical expression which agrees fairly well with the empirical observations. For a consideration of these theories a textbook on creep (2, 3) should be consulted.

Secondary Creep

The mechanism of the secondary, or steady-state creep stage is also largely still in doubt. To some extent a certain amount of slip may occur, but generally the higher the temperature at which creep takes place the more the flow becomes concentrated in the grain boundaries and sub-boundaries. The material contained between the crystals of a polycrystalline metal is conceived as an extremely thin disordered layer, possibly having the properties of a viscous material. However, it is found that the rate of flow observed during the secondary creep stage is dependent on the value of the applied stress to a greater extent than that associated with viscous flow. Furthermore, the steady-state creep stage is also observed in the study of creep in single crystals of metal, where no boundary material is present. There is a possibility, however, that the deformation occurs mainly in regions where slip has previously occurred, and where a large number of dislocations are present. Extending this concept to a polycrystalline material, it is thought that after the intergranular amorphous constituent has yielded sufficiently to cause the crystals to rotate slightly, interlocking of the crystal faces will occur. At such points, localised slip and movement of dislocations may occur and account for the additional deformation observed.

It is certain that this stage of creep is extremely sensitive to changes in the structure of the material and to any change taking place due to the temperature of the test. For this reason it is preferable to use a creep-resistant material in the condition most suitable at the service temperature rather than in the condition of maximum initial strength.

As will be seen later, the high-temperature alloy relies for its enhanced resistance to deformation on the precipitation of very hard particles within the grain boundaries. The presence of these particles serves to hinder grain boundary movements and reduces the creep rate accordingly. The precipitation is not usually complete before the metal is put into service, and subsequent precipitation occurs mainly during the secondary creep stage, with benefit to the alloy.

Tertiary Creep

The beginning of the tertiary creep stage is generally regarded as marking the end of the useful life of the alloy from the engineering view-point. This is due mainly to the unpredictable nature of this part of the creep curve. In some alloys the tertiary creep rate is only slightly greater than the steady-state creep rate, though increasing with time, and there may be a considerable interval between the end of the secondary creep and fracture, while with other alloys the onset of fracture may be very rapid.

Whereas few changes in the material are observable up to this point in the creep test, a small decrease in the density coincides with the onset of the tertiary creep stage and has been associated with slight changes in the metallographical structure of the alloy which gradually resolve into a form of crystal separation or micro-cracking.

There is no generally accepted explanation for this sudden change from steady creep. This is largely due to the fact that the type of failure frequently varies from one metal or alloy to another, and it may be that more than one mechanism is involved. Since a wide variety of materials have been examined, ranging from single crystals to complex precipitation-hardened alloys, and also since the properties of each will be influenced by the temperature and the rate of straining, it is not very surprising to find several modes of failure recorded.

Andrade (4) found that the tertiary stage of creep did not occur in the case of certain ductile metals if the constant stress system was substituted for the constant load. He suggested, therefore, that in the latter case, as the creep of the specimen gradually reduced the diameter of the test-piece, the applied stress increased and the rapid

tertiary creep was due to "necking" of the metal under such conditions.

This explanation cannot hold, however, when very little elongation of the specimen occurs. In addition, changes in the strain rate, at a point corresponding to the tertiary stage under tensile creep stresses, have been noted in compression creep testing (5), in which the dimensions of the specimen are changing in such a way that the applied stress is actually decreasing. The changes in strain rate are followed by a decrease in the creep rate and may be noted several times before total failure occurs.

In other cases fracture has been traced to the initial formation of surface cracks by oxidation or corrosion, or to structural changes in the alloy. Where failure cannot be related to either outside influences or dimensional alterations in the specimen, however, various theories have been put forward, but the explanations are insufficiently developed at the present time for these to be generally accepted.

Creep Testing

It will be appreciated that, since an entirely creep-resistant alloy is not possible, the gas turbine designer requires an accurate forecast of the behaviour of the various materials during service, so that the necessary allowances may be made for the change in the dimensions of the component. If, for example, the tolerance between the rotor blades and the casing was greater than necessary, there would be appreciable loss of efficiency, and if the gap was insufficient serious damage would result. The required allowance has frequently to be calculated by extrapolation from the results of creep tests over periods of only a few thousand hours. While this may be an undesirable feature of design data, the results of a few tests which have recently reached periods of the order of 100,000 hours have shown that for simple materials the extrapolation method is surprisingly accurate. To have real validity, a creep test should cover a period of at least 1000 hours and preferably longer where feasible.

Although there has been several methods of presenting the results of creep tests, a certain degree of standardisation has now been obtained, and the usual type of diagram is shown in Figure 22. This shows the relationship existing between the applied stress and the time required for different amounts of creep strain and rupture, at a given temperature. It is important to remember, however, that variations in the method of manufacture, composition, grain size, or heat treatment of an alloy will be reflected by a change in the position, slope and relative distances apart of the lines in such a diagram. The designer can determine from such diagrams the

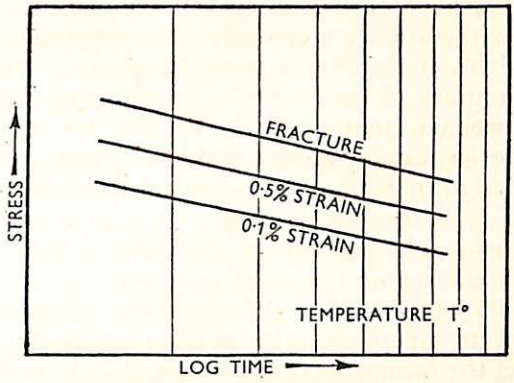


Fig. 22.—The arrangement generally used for presenting creep data at any one temperature

amount of stress at the chosen temperature which will produce a deformation not exceeding a specific rate, e.g. a rate of creep of 0·1% per 10,000 hours. Such a stress is known as the "limiting creep strength" at that temperature. The information from such a diagram can only be considered as a rough guide, since the results apply strictly to the conditions existing throughout the test, whereas the component will be subjected in service to several forms of stress, possibly under corrosive conditions. When comparing the creep properties of various materials, it is important that the comparison should be confined to the required service temperatures since it is possible to find an alloy with better creep properties than another at one temperature while the reverse may be the case at a different temperature. This is due to the fact that the temperature coefficient of creep is not the same for all alloys.

Measurement of Creep Rates

In considering the apparatus used in the measurement of creep rates it is sufficient in this book to refer only to the basic principles of all creep-testing machines, since a large number of different types are in use. Basically the machine must concentrate on three factors.

1. The steady application of a pure tensile load (or compressive in some cases).
2. Maintenance of a uniform temperature.
3. The precise measurement of the dimensional changes of the specimen when under load.

The tensile stress is generally applied through a lever arrangement, the fastenings to the specimen being passed through rigid guides to prevent any bending stresses being applied. The avoidance of any stress other than the axial tensile stress is most important in obtaining accurate creep measurements.

A wide variety of special temperature regulators are available for the furnace surrounding the specimen. Since the creep rate is highly susceptible to slight temperature differences it is vital to the accuracy of the results to prevent any significant variations of the specimen temperature. By the use of an efficient controlling device, usually mechanical or electronic, coupled with good furnace design, it is possible to control to within a fraction of a degree as long as may be required. It is sometimes advisable to arrange a metal tube around the specimen so that the thermal conductivity of the metal will assist in preventing any temperature variations along the length of the specimen. Fluctuations in the temperature of the creep-testing room from day to day would affect the control of the furnace temperature, and in a modern creep-testing laboratory considerable care is taken to maintain a uniform temperature.

For the most accurate results it is preferable to use the maximum specimen length possible, thus obtaining a more easily measurable extension for small strains, but the difficulties involved in obtaining a uniform temperature over a long specimen limit the gauge length which can be used in practice.

The accurate measurement of the very small changes in length associated with creep-testing calls for an extensometer of high sensitivity, yet having sufficient strength and stability to maintain its accuracy throughout the duration of the creep test. The most frequently used instrument is the Martens extensometer. This consists essentially of two rods passing into the furnace alongside the specimen, one rod being firmly attached to the specimen at each end of the gauge length. The ends of the rods protruding from the furnace are held together by springs, but slightly separated by a rhomboid-shaped piece of steel to which a small mirror arm is attached. As extension of the specimen occurs the two rods move relative to each other and twist the rhomboid. The deflection of the mirror may be measured with a suitable light beam and scale.

The frequency of the measurements will depend on the sensitivity of the extensometer and the rate of extension of the specimen. In commercial routine creep-testing where a long time is impractical it is usual to use loads such that the creep test is concluded in a matter of days. In such cases the readings may be taken hourly. The more elaborate tests connected with the theoretical aspect of creep, however, last for several months or even years and the readings will therefore be taken at more infrequent intervals. It is generally found that the longer the test period the more reliable the interpretation of the results becomes, especially in their application to design requirements. For short-time tests several special methods have been developed, such as the Hatfield Time-yield, the Barr

and Bardgett and the Harris tests. Full descriptions of the methods are to be found in books devoted to creep and creep-testing.

Some of the factors which cause changes in the creep properties of the alloy during service will now be considered.

Metallurgical Factors influencing Creep Properties

A good deal of the research work concerned with the development of creep-resistant alloys has of necessity been of an *ad hoc* nature. Until the fundamental principles of the creep mechanism are more fully understood, however, this approach is the only one available, despite the considerable amount of work involved. From a consideration of the process it is evident that there are two approaches to improvement of the creep properties of an alloy, either to increase the stiffness at the operating temperature or to slow down the rate of the processes leading to failure.

It is well known that certain metals possess better inherent high-temperature strength than others, and such metals would be preferred as the basis of the alloy. Unfortunately some metals with very good high-temperature properties, such as for example chromium and molybdenum, have other serious disadvantages and the present alloys are all based on the metals iron, nickel and cobalt. The enhancing of the inherent resistance to deformation of these metals to the much greater values required, without serious loss of ductility, is brought about by a process known as "precipitation hardening" or "age-hardening". Consider a binary alloy represented by the line AB in Figure 23. If this alloy is heated to

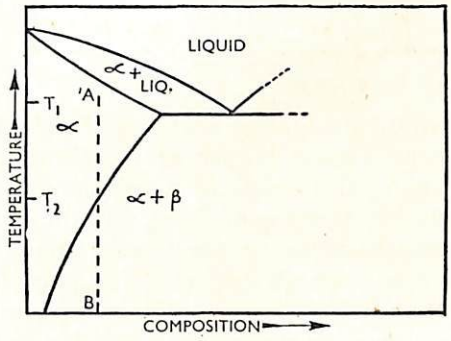


Fig. 23.—A simple equilibrium diagram for a binary alloy system in which precipitation hardening may be produced

temperature T_1 there will be only one phase present, the α-phase. On slow cooling, the α-phase will become unstable below temperature T_2, and a certain amount of the β constituent will appear as a precipitate throughout the alloy. Usually, under these conditions, this process will not greatly alter the hardness of the alloy. If, however, instead of slow cooling, the alloy is quenched rapidly

from temperature T_1, the formation of the β-phase will be effectively prevented. The resulting material will be thermodynamically unstable and there will be a tendency for the β-phase to come out of solution on standing. In certain alloys this may occur at room temperature and if the hardness of the alloy is followed over a period of time from the moment of quenching, a curve of the type shown in Figure 24 is obtained. The increase in hardness is associated with the nucleation of the β-phase particles, the hardness falling again as the particles become discernible as a separate entity. If the "age-hardening", as this phenomenon is called, is now repeated at higher temperatures it is found that the maximum hardness is reached in a shorter period of time but the actual value of the maximum hardness is lowered as the temperature increases. Despite the fact that the hardness falls appreciably with time after the maximum has been reached, the mechanical properties of the alloy are greatly improved, due to the presence of the finely dispersed second phase along the grain boundaries and slip planes.

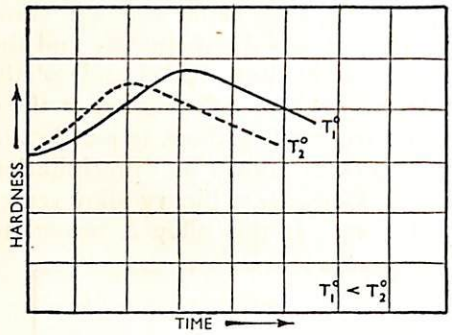


Fig. 24.—The effect of time and temperature on the hardness changes taking place during age-hardening

A well-known alloy of the age-hardening type is the aluminium-4% copper alloy known as Duralumin. The heat treatment for this material consists of a solution treatment at 500°C (935°F) followed by water quenching. In the course of from four to five days at room temperature the tensile strength increases from about 18 ton/in^2 to some 26 ton/in^2. The particles precipitated in this case are of $CuAl_2$, an intermetallic compound.

In a modern creep-resistant alloy, where many hardening elements are present, it is not always known for certain what the dispersed phase actually contains, but, in general complex mixed carbides, nitrides and silicides of such metals as tungsten, vanadium, molybdenum, columbium, titanium, etc., are most frequently found, together with intermetallic compounds, such as the columbides, molybdides and tungstides of iron. Such compounds are taken into

Plate 10. *Precision contour milling a compressor blade*

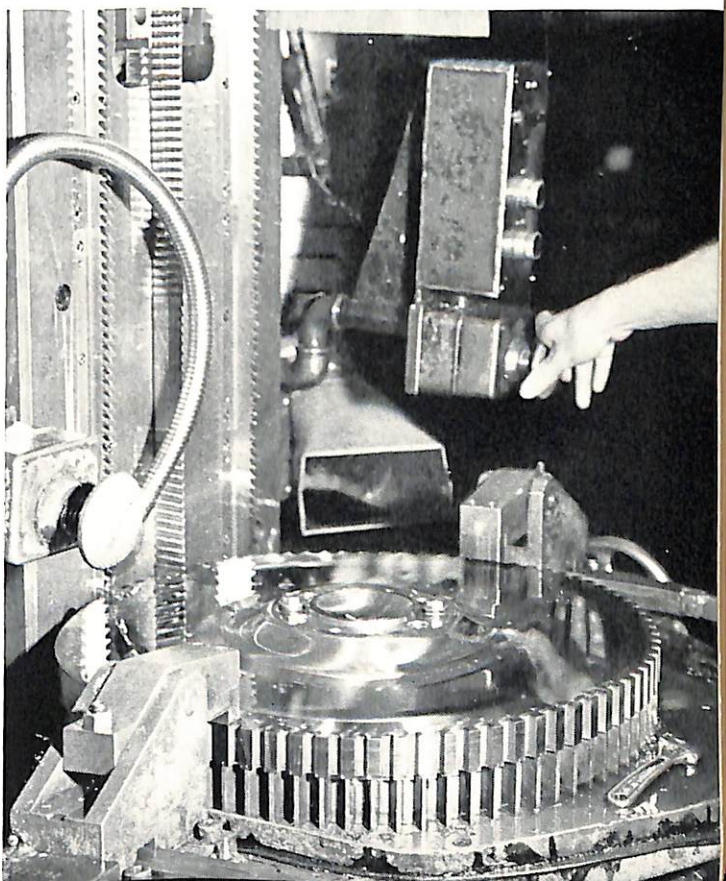


Plate 11. *Broaching a turbine disc*

Plate 12. *Drop forging a turbine disc for an aircraft gas turbine*

solution at temperatures of 1000–1300°C (1830–2375°F), and subsequently reprecipitated by heating at 700–800°C (1300–1475°F). It is important to note that a heat-treatment giving maximum creep resistance at a certain temperature may not be the most suitable for a different temperature.

The development of the maximum creep-resisting properties will depend on the efficiency of the solution treatment which will be governed by the time and the temperature of treatment and, in turn, these will be restricted by the surface deterioration tolerable when holding the alloy at a very high temperature in a furnace atmosphere. Since the heat-treatment must be carried out after the primary forming operations, it is preferable to avoid a severe heat-treatment, even at the expense of complete solution, to reduce the final surface cleaning to a minimum.

The optimum precipitation treatment may be readily ascertained by laboratory tests, but since even small variations from the required temperature and time may result in a serious discrepancy between the properties obtained and those required, it is essential that the large-scale production should be closely controlled.

In addition to the hardening brought about by precipitation, some of the constituents of the alloy remain dissolved in the matrix and contribute to both the stiffening of the alloy and the retardation of thermal softening. The addition of a solute element invariably increases the yield strength, tensile strength and hardness of the solvent metal. Although this is known to be caused mainly by the strain induced in the lattice by the solution of the foreign atoms, this is almost certainly not the only mechanism involved, since several examples are known of solution hardening occurring without any detectable lattice distortion. It is generally true that the most effective solute elements for increased creep resistance are those which contribute to the stiffness without appreciably lowering the melting point of the solvent metal.

It is sometimes stated that, at high temperatures, the creep properties of coarse-grained alloys are superior to those of a similar alloy with a fine grain size, the reverse effect being the case at lower temperatures. This is based on the conception of a metal as a number of crystals surrounded by a thin disordered boundary constituent, the latter having properties similar to those of a viscous substance. At elevated temperatures, where the creep occurs to a large extent by deformation of the boundary material, the larger the grain size the less the deformation should be for a given stress. The experimental evidence on the effect of grain size on creep properties tends to support this argument but since very coarse-grained material has rather inferior mechanical properties, any additional

creep resistance so obtained is more than outweighed by the increased brittleness.

Stress-rupture Testing

The stress-rupture, or creep-rupture, test has been widely used, particularly in America, for assessing the relative merits of various alloys for high-temperature service. The test is similar to the normal creep test, but it concentrates entirely on the time required for fracture under a range of stresses. The ordinary high-temperature tensile test is a form of stress-rupture test in which the ultimate tensile stress causes fracture in a very brief space of time. When lower initial steady stresses are used the time to fracture becomes progressively longer. From the results a diagram of the type shown

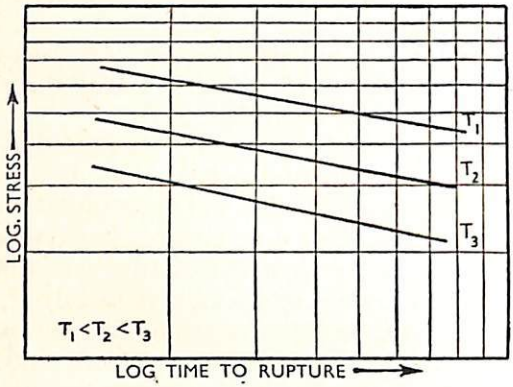


Fig. 25.—A typical plot of stress-rupture test results at several temperature levels

in Figure 25 is plotted, and the lines may be extrapolated in a similar manner to the creep curves. In stress-rupture diagrams plotted on a log-log basis a break in the slope of the line is frequently observed, though the time at which this occurs is usually only of the order of tens of hours and apparently does not affect the validity of the results in tests at low stress levels. It is thought that this break coincides with the change-over from trans-granular to inter-granular fracture of the test-piece, since trans-granular fracture is usually associated with short-time failures, while inter-granular fracture is more liable to occur the higher the temperature and the lower the rate of straining.

It is interesting to note that this break in the curve is not observed in non-oxidising conditions and becomes more pronounced as the oxidising power of the atmosphere is increased. This effect has been connected with the observation that inter-granular fractures begin in the surface layer of the specimen and there is little

doubt that after the start of fine surface cracks the oxidation of the exposed metal accelerates the general failure.

In general the stress-rupture test requires less elaborate apparatus than a creep test, the sole requirements being an accurately controlled furnace and some arrangement for applying the load. However, the information obtained is much more limited. It is desirable to record the elongation observed after failure, to indicate the extent to which the alloy undergoes loss of ductility on prolonged heating.

It must be pointed out that the measure of strength obtained by stress-rupture testing is not necessarily related directly to measurements of the stress required to produce specified creep rates, i.e. the rate of the second stage of creep. In designing for high-temperature service, however, the information obtained from both tests is supplementary, since it is essential when extrapolating creep curves on a time basis to be certain that fracture will not supervene.

A special type of test in which the stress is gradually reduced as creep takes place may be mentioned here. This is known as "relaxation testing" and simulates the service conditions which will be experienced by materials such as those used for bolts. When a tightly bolted part is put into service the initial elastic strain in the bolt is slowly changed to plastic strain as creep takes place, the applied stress being slowly reduced in consequence.

Fluctuating Stress

Although the fluctuating stresses set up in the gas turbine are considerably less severe than in a combustion engine, a certain amount of aerodynamic vibration is set up by the beat of the compressor blades against the air being compressed, and this is transmitted through the combustion chamber, possibly affecting the turbine blading. From the design aspect it is interesting to note that this vibration is less marked when an axial flow compressor is used. In the turbo-propeller gas turbine vibrations which may be transmitted to the rest of the engine are also set up in the propeller and reduction gear.

Consequently the materials of the gas turbine are subjected to rapidly alternating stresses of small amplitude and the effect on the construction materials may result in failure by "fatigue". If the number of stress reversals required to cause fracture at room temperature over a wide range of stress values are determined, a curve similar to that in Figure 26 (*overleaf*) is obtained. It will be noted that below a certain stress value the life of the test piece becomes infinite. For steels, stresses which do not cause failure after 10 million cycles are regarded as being below the fatigue limit. In

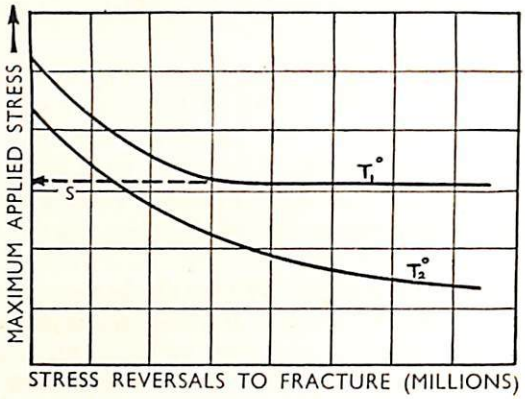


Fig. 26.—Typical fatigue curves. The tests at low temperatures ($T_1°$) show a maximum safe stress level (S), which disappears with increasing temperature

the case of a given steel, the stronger it is made by heat-treatment the higher will be the limiting fatigue stress. Along with the increase in strength, however, the sensitivity to stress concentration may also become greater, so both factors must be balanced in designing a component subjected to fluctuating stresses. The stresses used in these tests are substantially lower than those used in static stress tests and the behaviour of an alloy in a fatigue test cannot be predicted from its behaviour under steady stresses. At elevated temperatures, it is found that the curve does not become horizontal but continues to decrease slowly.

The occurrence of a fatigue failure at low temperatures generally results from work- or strain-hardening, which reduces the ductility of the material to such an extent that a crack will eventually form at some weak point, usually a surface irregularity, since the stresses are highest at the surface. The crack will be propagated across the section at a rate which will be dependent on the stress range. Eventually the amount of material maintaining the load becomes so small that fracture occurs by simple overloading. Consequently a fatigue fracture is characterised by a smooth outer rim with, in some cases, concentric lines around the point of origin of the fracture, and a rough central region showing the normal fracture of the material.

The surface defect initiating the fatigue failure is usually very slight, e.g. a machining groove or a minute crack produced during heat-treatment. To reduce the tendency for this type of fracture to occur, therefore, the components subjected to fluctuating stresses must be treated by a high-quality surface finishing process. Other areas of stress concentration include non-metallic inclusions, blowholes and sharp changes of section. Inspection by the modern methods for crack-detection, supersonic, magnetic, electrostatic,

etc., ensures that flaws introduced during manufacture are quickly detected.

In fatigue testing at elevated temperatures similar conditions apply, but, in addition, the factors leading to fatigue failure are more complex and the temperature and frequency of the stress cycle have to be considered. Research on this aspect of fatigue has not been so extensive as the studies of creep resistance but it is known that there is no obvious relationship between the static rupture stresses and the values of the cyclical stresses. In general, however, the fatigue properties improve with the creep resistance.

A limited amount of work has been carried out to determine the influence of fluctuating stresses superimposed on a steady tensile stress. The published results are interesting in that they indicate that the importance of a cyclic stress in accelerating creep decreases with increasing temperature. At the higher temperatures a fluctuating stress may even decrease the creep rate due to a steady tensile stress, though for such an effect the maximum cyclic stress value required is probably much higher than is encountered in service.

In practice the effects of fluctuating stresses can be greatly minimised by careful designing. The damping capacity of the majority of gas turbine alloys is low and this must be supplemented by using a construction giving the minimum stress production and transmission. The limited life of certain early designs of combustion chamber was partly due to the vibrations set up by the air stream causing failure in the spot welds. This is now avoided by the use of deep-drawn material. Similarly in the case of centrifugal compressors it is common design practice to have unequal numbers of impeller and diffuser vanes, and to have a prime number of vanes in at least one of these components. This ensures that, at any given instant, only one impeller vane is passing a diffuser vane, thus reducing the induced impulses. In axial compressors, the blades are fixed to the disc by a loose arrangement which increases the stress damping.

Shock Stresses

Increasing the stiffness of an alloy eventually results in such a loss of ductility that any sudden shock may cause fracture. This condition will be aggravated where a sharp change of cross-section occurs. Since the components of a prime mover should be reasonably robust, it is essential that the material should have a fairly high impact strength.

In addition to mechanical shocks there are also shocks of a thermal nature. Thermal stresses arise when there is a steep temperature gradient set up in the material and the properties of the material

which will influence the value of such stresses are the modulus of elasticity, the thermal expansion coefficient and the thermal conductivity. During the starting up or shutting down of a gas turbine severe temperature gradients will be set up and the high thermal stresses may lead to fracture. Failures from these causes are not unknown. If the fracture occurs within the course of a few cycles the failure is referred to as "thermal shock", whereas if a large number of cycles are required, the material is considered to have failed by "thermal fatigue". The mode of failure in each case is similar to that occurring when the stresses arise from mechanical causes.

In addition to the mechanical properties of the material the stability of the metallographic structure of the alloy must be studied, and also its behaviour when subjected to the corrosive action of the products of combustion of the fuel.

Structural Stability

It has been shown above that the properties of a creep-resistant alloy depend on the existence of a particular structural condition, i.e. the presence of a finely dispersed constituent along the grain boundaries and slip planes, together with a strong matrix. During service at elevated temperatures, changes in this structure are almost inevitable. The changes which may occur are:

(a) Alteration of the size, composition or degree of dispersion of the precipitated particles.
(b) Precipitation of some new phase or a phase change.
(c) Recrystallisation.

(a) The effect of diffusion on finely dispersed particles in an alloy results in agglomeration and spheroidisation. This decreases the stiffness of the material. Fortunately the complex carbides, which generally predominate, are very slow to diffuse even at temperatures as high as 700 °C (1300 °F) and the presence of elements such as nickel and chromium in the matrix reduces the diffusivity still further.

(b) 1. In the case of iron-base alloys operating at temperatures of the order of 600–700 °C (1100–1300 °F) there is a possibility that, over long periods of time, the brittle sigma phase of the iron-chromium system may be formed. This is an intermetallic compound, nominally consisting of about 50% iron and 50% chromium, and under certain conditions its formation may result in the alloy becoming extremely sensitive to shock, particularly at room temperatures. The range of alloy composition in which it may occur is rather wide at 600 °C (1100 °F), being found in alloys with a chromium content at least as low as 20%, but narrows with increas-

ing temperature until it disappears between 800–900°C (1475–1650°F). The presence of other elements usually influences the composition range to some extent, e.g. both manganese and silicon broaden the range.

In the last five years new surveys of the binary alloys of the transition elements have shown that phases isomorphous with the iron-chromium sigma phase are to be found in many cases. At the present time such phases have been reported in the following systems:

(*i*) Cobalt-chromium
(*ii*) Iron-vanadium
(*iii*) Iron-molybdenum
(*iv*) Nickel-vanadium
(*v*) Cobalt-vanadium
(*vi*) Chromium-manganese
(*vii*) Vanadium-manganese

In the iron-molybdenum system the Fe-Mo phase is unusual in that, unlike the other systems, it is only stable above 1180°C (2156°F). Below this temperature it breaks down to α-molybdenum and the ϵ-Fe_3Mo_2 phase.

Sully (6) has suggested that the appearance of the sigma phase can be explained if it may be assumed that the incidence of the phase is related to a critical excess of electrons in bond orbitals over vacancies in the 3d atomic orbitals. On this assumption calculations of the composition limits of the phase in the various systems agree well with the known experimental results.

It will be realised, therefore, that many commercial alloys have compositions which make the formation of the sigma constituent in service possible. The alloys of the 18% chromium-8% nickel, 25% chromium-20% nickel, 25% chromium-12% nickel and the 25% chromium types are in this category. In general, however, the amount of sigma formed is small. To some extent the precipitation of this hard constituent could be beneficial, if it occurred as a fine dispersion, but the brittleness results from its tendency to appear in massive form at the grain boundaries.

2. When austenitic stainless steels of the chromium-nickel type are heated for some time in the temperature range 450–875°C (850–1600°F) precipitation of chromium-rich carbides occurs at the grain boundaries. This precipitation renders the alloy susceptible to inter-granular corrosion, since the carbide formation is related to a decrease in the chromium content of the alloy in the region immediately adjacent to the carbide particles.

This type of corrosion has been called "weld-decay" from its tendency to occur close to welded joints. Since, in making a weld, the temperature varies from the melting point of the metal to room

temperature over a range of a few inches, some part of the metal will be heated to the dangerous range. Consequently corrosion occurs not at the weld but a short distance from it.

The carbide formed is rich in chromium because of the high affinity of chromium for carbon and one method of preventing weld decay involves the addition of much stronger carbide forming elements, such as titanium and columbium, to combine preferentially with the carbon and prevent loss of chromium from the solid solution. In practice it is usually found necessary to add several times the amount of titanium or columbium required for combination to ensure that all the carbon will be accounted for.

An alternative method consists of heating the steel at a sufficiently high temperature for the precipitation to occur but also allowing adequate time for the chromium depreciation to be eliminated by diffusion. Both processes are referred to as "stabilisation".

Other possible preventive measures include a heat-treatment after welding, to take the carbides back into solution again, followed by quenching, and also the production of steel with a very low carbon content. The first is, however, impracticable with most welded constructions since it requires the use of large furnaces and the second is difficult to achieve in practice.

3. The carbon-molybdenum steels had been in extensive use for some years when a serious explosion in 1943 showed that after prolonged service at temperatures greater than 450°C (850°F), graphitisation could occur in such a way as to produce lines of weakness adjacent to welds. It was subsequently found that this process is accentuated by the presence of aluminium, but retarded by additions of molybdenum, chromium and vanadium, according to the amount present. Since these discoveries greater care has been taken in selecting low-alloy steels for use at such temperatures and higher chromium, molybdenum or vanadium contents are now used than would otherwise have been the case.

(c) Recrystallisation occurs when cold-worked material is heated above a certain critical temperature, typical of the material and to some extent dependent on the amount of deformation applied to the particular material. It involves the formation of new, unstrained crystals from the stressed ones and consequently a return to the lower mechanical properties of the unstressed material, i.e. decrease of strength and increased ductility. If the deformation occurs at the high-temperature recrystallisation would occur almost simultaneously and the continued application of stress would cause continuous deformation. Fortunately, the low strain rates involved in the creep of an alloy are usually insufficient to cause extensive recrystallisation at the present operating temperatures.

Manufacturing Variables

In addition to the variation of creep and fatigue properties possible by composition changes, it is recognised that otherwise identical alloys may have differing properties if the manufacturing process is altered. Consequently alloy manufacturers have carried out a number of investigations to elucidate the effect of such variables as melting and de-oxidising practice, casting temperatures, ingot design, etc. The number of inclusions in the steel, for example, should clearly be kept to a minimum since such particles would serve as stress raisers and crack propagators.

Oxidation and Corrosion

The highly oxidising conditions which exist in the combustion chamber and turbine call for a considerable resistance to oxidation on the part of the alloys concerned. The behaviour of metals and alloys when exposed to air or oxygen at elevated temperatures has been extensively studied for a number of years and the mechanism of oxide formation is fairly well understood.

When oxygen comes in contact with a clean metal surface, a thin unimolecular layer of oxide can form by chemical combination.

The further course of the oxidation process then depends on the nature of the oxide and metal concerned. Thus if the oxide occupies a smaller volume than the metal it replaces, a porous oxide layer results and further access of molecular oxygen to the underlying metal is possible. This will proceed until all the metal is

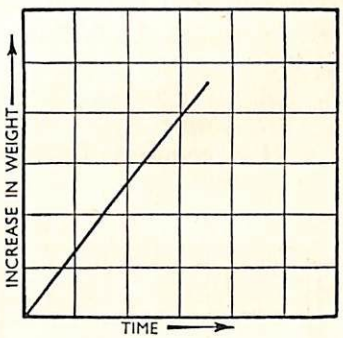


Fig. 27.—The linear oxidation curve obtained when diffusion is not a limiting factor

consumed and, since there are no restraining factors, the reaction will show a linear relationship with time, as shown in Figure 27, i.e.

$$\frac{dw}{dt} = k$$

or

$$w = kt + A$$

where w = weight of oxide formed at time t, and k and A are

constants, A being zero if the experiment starts with an oxide-free surface. This is the linear oxidation law and is obeyed by such metals as calcium, magnesium and other alkaline earth and alkali metals.

In the case of a metal whose oxide occupies a greater volume than the metal from which it is formed, then the oxide should be relatively non-porous and if further oxide is to be formed then diffusion of metal or oxygen through the oxide layer has to occur.

It has been known for a long time that the metal and oxygen atoms do not diffuse as such, but split up into ions and electrons, since these have much smaller radii than the atoms. The mechanism envisaged for the diffusion of metal to the surface to form an oxide unit is outlined below. The metal atom first splits up into an ion and an electron. The electron travels rapidly to the surface of the oxide and ionises one of the oxygen atoms adsorbed there. The metal ion follows, at a much slower rate because of its greater size, down the concentration gradient and, on arrival at the metal oxide surface, it combines with the oxygen ion to form a molecule of oxide.

$$M \rightarrow M^+ + e \qquad e + O \rightarrow O^- + M^+ \rightarrow MO$$

A similar picture may be used to represent the inward diffusion of oxygen.

Since the concentration of oxygen is a maximum at the oxide-gas surface and a minimum at the oxide-metal interface, while the reverse is true for the metal concentration, the rate of diffusion over this concentration gradient is proportional to the length of the diffusion path, i.e. the thickness of the oxide layer.

Then, if w = quantity of oxygen,
h = thickness of the oxide layer,
t = time
and k = constant of proportionality

then $$\frac{dw}{dt} \propto \frac{1}{h}$$

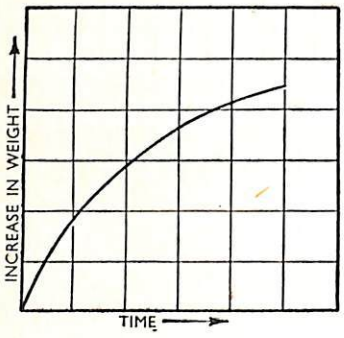


Fig. 28.—The parabolic oxidation curve obtained when diffusion within the scale layer controls the rate of oxidation

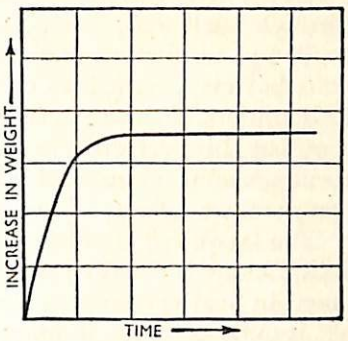


Fig. 29.—The asymptotic curve obtained with oxidation resistant materials

and since all the diffusing oxygen further increases the thickness of the oxide layer,

$$w \propto h$$

or

$$w = k_1 h$$

$$\therefore \frac{dw}{dt} = \frac{k_1}{w}$$

Integrating

$$w\,dw = k_1\,dt$$

or

$$w^2 = kt$$

the integration constant again being zero for an initially clean metal surface.

This is the parabolic oxidation law (Figure 28) and is followed by a large number of metals such as iron, cobalt, nickel, etc.

The most important oxidation process in the development of high-temperature alloys, however, is associated with only a few special oxides, such as Cr_2O_3, Al_2O_3 and BeO. By virtue of their ionic structure these oxides do not allow the diffusion of either metal or oxygen ions through themselves once a complete film has been built up, the thickness of the film varying according to the temperature concerned. Consequently the weight increase versus time curve for such a process is of the type shown in Figure 29. This is usually referred to as asymptotic oxidation.

When an alloy is oxidised, each constituent contributes to the oxide layer, although not necessarily in the same proportion as the metals exist in the alloy. The constitution of the oxide layer is dependent on the relative rates of diffusion of the metals in the alloy, and most important, on the affinity of each metal for oxygen. For example, if an iron-base alloy containing 20% chromium is heated at temperatures of 600–700°C (1100–1300°F) it is found that the scale formed consists almost entirely of chromic oxide. This

chromic oxide layer has the property of inhibiting the movement of ions through itself and, consequently, once a complete layer has been built up, no further oxide can form. While the layer is incomplete, however, some iron oxide can form and this accounts for the small amount of this material found in the scale. Several other oxides possess this protective property, as mentioned above, and the two elements chromium and aluminium are always present in high-temperature alloys, either separately or in combination to supply the required oxidation resistance. While the use of aluminium alone is usually restricted to copper-base alloys, such as those used in heat exchangers, the use of chromium to the extent of 15–30% is quite common in other alloys for operation up to 1100°C (2000°F). Chromium cannot be used in copper-base alloys because of its very low solubility. The chromium generally contributes little to the creep strength of an alloy, its main function being to provide the essential protection against the oxidising atmosphere.

It will be realised that once a protective oxide layer has been built up on the surface of a high-temperature component, it is highly desirable that the layer should remain firmly attached to the metal throughout its service. If this were not so, and the oxide became detached frequently, the alloy would undergo appreciable dimensional changes. Since an oxide does not have the same thermal expansion coefficient as the underlying metal it will be appreciated that as the metal is heated or cooled the scale and the metal will not expand and contract at the same rates. As a result, stresses will be set up at the scale-metal interface. If the stresses are high and the scale fairly rigid in character, flaking away of the scale under thermal fluctuations is almost certain to occur. Consequently it is desirable that the protective oxide film formed on a heat-resistant alloy should be able to withstand a reasonable amount of temperature fluctuation.

In the gas turbine, in addition to air, there are the products of combustion of the fuel to consider. The possibility of corrosion arising from this source depends largely on the fuel used. In the case of the aircraft turbine where paraffin is used, no trouble is experienced since with air-to-fuel ratios of the order of 100:1 the concentration of the carbon dioxide, water vapour and sulphur dioxide produced is negligible.

In land-based or marine gas turbines, however, paraffin is too expensive for economic running, and it is desirable to use such fuels as residual oils, low-grade coals, peat or other cheaper materials. The products of combustion of these materials contain larger amounts of corrosive substances, which may lead to loss of the protective oxide coating and damage to the underlying metal.

The most important corrosive processes encountered are discussed below:

Vanadium Pentoxide

Crude petroleum contains a large number of metallic compounds of a very complex nature, incorporated from the vegetable matter from which the oil was formed. When the oil is burnt a large number of metal oxides are formed from these compounds and these comprise the main ash constituents. During the distillation process many of the volatile fractions are almost freed from these ash-forming constituents but the heavy fuel oils, with which a proportion of distillation residues are blended to form the so-called "residual" fuels, form from 0·01% to 0·1% ash on combustion. Of the many metal oxides found in these ashes the most important has proved to be vanadium pentoxide. This oxide may comprise up to 70–80% of the ash from certain oil-fields while in others it may be almost absent.

The severe corrosive action of vanadium pentoxide was first observed in experimental turbines using heavy fuel oils when blade temperatures approached 650°C (1200°F). Experiments have shown that the action of the vanadium is greatly accentuated by the presence of certain metallic salts such as sodium sulphate, formed from the sodium compounds and the sulphurous gases present in the products of combustion, and metal chlorides.

The melting point of vanadium pentoxide is only 673°C (1243°F) and this is further lowered in the presence of other compounds, such as metal vanadates. Consequently an ash which contains a large proportion of vanadium pentoxide will be either fluid or semi-fluid at the turbine temperatures and the hot ash particles will tend to stick to any metal surface on to which they impinge. At these points severe pitting and corrosion of the metal occurs.

On investigation into the effects of vanadium pentoxide on a variety of metals and alloys, especially the heat-resisting types, it was soon found that the fluxing effect, combined with the high oxidation potential of the oxide, almost justified the use of the term "universal solvent". No metal or alloy has so far been found which can withstand prolonged contact with vanadium pentoxide without ultimate deterioration. Despite intensive investigation by several research establishments no generally accepted mechanism of the attack is forthcoming although the general principles are known.

From what has been said earlier in the chapter it will be remembered that whereas the majority of metals form oxides which allow diffusion of metal or oxygen ions and electrons through them,

chromic oxide is unusual in that the diffusion process stops completely once a certain thickness of oxide has been built up. Consequently it follows that should this oxide film be interfered with in some way so as to break down this property, diffusion will recommence and the underlying alloy, deprived of its protection, subjected to prolonged oxidation.

When vanadium pentoxide is brought into contact with pure chromium or a heat-resistant alloy, protected by a chromic oxide film, at temperatures above about 500°C (935°F) vanadium ions probably enter into solid solution in the chromic oxide in such a way that the diffusion limiting factor is overcome and oxidation proceeds unchecked. This process is possible even when the vanadium pentoxide is present in the solid state, since even at such temperatures the vanadium readily enters the oxide lattice. If the vanadium pentoxide is molten, then direct fluxing of the oxide will occur and oxidation of the metal proceed very rapidly until sufficient oxide has formed to solidify the mixture. The oxygen for this reaction probably comes, in the first place, directly from the vanadium pentoxide, since above the melting point this compound is partially dissociated—

$$V_2O_5 \rightarrow V_2O_4 + (O)$$

but the removal of any oxygen would be counteracted by further solution at the liquid-air interface.

Once a solid scale has been built up the reaction must proceed by solid state diffusion unless a porous scale is formed. The observed reaction rates are very great but not linear with time so that diffusion is probably the most important factor, and the rate of diffusion is very rapid.

From the practical point of view, however, the main interest centres around the early stages of the process since a continual supply of vanadium pentoxide is being carried in the ash and deposited on the blade surface. The concentration of vanadium in the scale consequently remains very high, according to the ash content of the fuel and the rate of combustion. Under such conditions the rate of corrosion is very high indeed.

Research work has shown that the chances of developing a commercial alloy capable of completely withstanding the attack of molten vanadium pentoxide appear to be rather remote and consequently the experiments have been directed towards either removing the vanadium compounds from the oil before combustion or introducing some substance into the burning fuel stream to combine with the molten ash and render it innocuous before it reaches the turbine blades.

The attempts to remove the metallic compounds from the fuel have so far proved mainly unsatisfactory. Little is known about the nature of the organo-metallic components in the oil and it is partly due to lack of this information that extraction methods are difficult to devise. Various solvent techniques have been tried but none has so far been found to extract the required compounds without considerable wastage of oil. At the present time the only satisfactory pre-treatment known involves distillation, which adds appreciably to the cost of the fuel and removes the economic advantage of the residual oils.

The alternative "oil additive" technique has also been thoroughly explored. By laboratory and rig tests it has been shown that the stickiness and reactivity of the ash can be overcome by the addition of basic oxides, such as those of magnesium and zinc, into the burning fuel. The addition may be made either directly as the oxide or as an oil-soluble compound which will form oxide along with the vanadium compounds during combustion. The latter method is extremely expensive, however, and can almost certainly be ruled out for practical purposes. Unfortunately the amount of additive required may be as much as 10 times the weight of the fuel ash and the powdery deposits formed introduce problems of blade fouling and choking of the heat-exchanger tubing. With specially designed blade shapes, deposition on these surfaces could probably be avoided, but it would seem desirable to introduce some form of gas cleaning before the heat-exchanger.

The most important problem concerned with the use of additives is connected with the sulphur content of the fuel. The basic oxides, while combining with the ash, also form sulphates with the sulphurous gases present. In general it seems that the tendency of most of the suitable oxides to form sulphates is higher than that to form vanadates. Since the sulphur content of the residual oils is frequently greater than 1%, a great deal of the value of the additive may be lost in this way. Whether or not the physical effect of drying up the ash particles irrespective of the actual composition will be sufficient to minimise blade corrosion is still in doubt.

An alternative additive proposal, that of adding some substance which will combine with the vanadium to form some highly volatile compound, appears to have made little progress, probably due to the fact that inorganic volatile compounds are not common and organic ones would be unstable at such temperatures.

Some success in reducing corrosion problems has been achieved by controlling the combustion of the fuel so that a small proportion of unburnt soot particles are formed. These tend to surround the

ash particles and prevent them sticking to the metal. The loss of combustion efficiency required is said to be less than 1%.

Whichever method is finally selected one thing appears certain. Unless the cost of the final fuel and its treatment is substantially below that of the heavy distillate oils, the present trend towards the adoption of the gas turbine in place of the Diesel-type engine will be adversely affected, since only the use of cheaper fuel can compensate for its higher fuel consumption at the present operating temperatures.

Sulphur

The corrosive effects of sulphur-bearing gases on steels and other heat-resisting alloys have been extensively studied. In general, if the atmosphere is highly oxidising the presence of the sulphur has no influence on the rate of oxidation of the alloys. If, however, the oxygen content of the atmosphere falls below about 5% then serious corrosion (7) may occur, particularly in the case of nickel-rich alloys. This is due to the formation of low-melting-point sulphides or oxy-sulphides in the scale layers. Nickel sulphide is particularly bad, since its melting point may be as low as 620°C (1148°F). In nickel-rich alloys, corrosion involving sulphur gives rise to a characteristic green product from which the term "green rot" has arisen.

It might be supposed that in the gas turbine, where the air to fuel ratio is so high, reducing or weakly oxidising conditions could not arise, but in turbines burning solid fuels such as coal and peat, with high sulphur contents, it is possible to have partially burnt fuel particles smouldering on metal surfaces and setting up locally reducing atmospheres. In experimental turbines some corrosion has been observed in air pre-heater units where ash deposits have set up such conditions. This trouble can, however, be prevented by careful attention to the complete combustion requirements of the granular fuel.

One other form of corrosion which may be traced to the sulphurous gases may occur in equipment such as heat-exchangers when the turbine is shut down for a short time. It is found that the cooling of the gases results in the formation of sulphuric acid droplets on the metal surfaces, corrosion occurring when evaporation concentrates the acid to the required strength. In normal working such conditions should not arise since the temperature of the gases would not fall below the dew point.

Catastrophic Oxidation

A very unusual type of rapid oxidation which has been observed in the last few years is associated with the use of alloys containing

appreciable amounts of molybdenum. This metal forms an oxide MoO_3 which melts at 795°C (1460°F) and is extremely volatile at temperatures from about 600°C (1100°F) upwards. It was noticed (8) that, under certain conditions in the heat-treatment furnaces, alloys containing more than 4–5% molybdenum are liable to oxidise completely away in a very short time. This was subsequently found to occur only when the atmosphere surrounding the alloy was completely stationary, and it was suggested that under such conditions very high concentrations of MoO_3 could be built up in the scale layers. This destroyed the protective film effect, possibly by the formation of molybdates (9) molten at such temperatures.

The effect was also noticed in certain parts of operating turbines in which these alloys were being used, such as the fir-tree root fixings, where there are small openings in which the air circulation is very poor.

The rapid oxidation can be completely prevented by ensuring adequate air circulation around the metal, so that any molybdic oxide is removed by volatilisation as fast as it is formed and the protective film allowed to build up without interference.

Erosion

Where a rapidly moving gas stream impinges on a metal surface, the presence of even very minute solid particles in the gas may result in rapid abrasion of the metal. In a steam turbine for example it is essential to avoid the use of steam containing the slightest trace of water droplets since otherwise rapid erosion of the turbine blading occurs. In the use of solid fuels in a gas turbine, small particles of ash will be carried over from the combustion chamber, unless special gas-cleaning apparatus is installed, and such particles may be particularly erosive.

The degree of gas cleaning required still remains doubtful and may vary from one turbine to another according to the nature of the ash formed and the design of the turbine. American experiments with a gas turbine intended for locomotive use indicated that particles less than 20 microns may be tolerated but above this limit rapid erosion of the blading could occur.

This type of damage, together with the corrosion and fouling problems arising with the use of solid fuels, undoubtedly make the use of the closed-cycle gas turbine very attractive, although the air pre-heater then presents similar problems.

REFERENCES

1. E. N. da C. Andrade. *Proc. Roy. Soc.* (London), Vol. A.84, p. 1, 1910; Vol. A.90, p. 329, 1914.
2. L. Rotherham. *Creep of Metals*, Institute of Physics, London, 1951.
3. A. H. Sully. *Metallic Creep*, Butterworths Scientific Publications, 1949.
4. E. N. da C. Andrade. *Proc. Bristol Conference on Strength of Solids* (London, Physical Soc., 1948).
5. A. H. Sully, G. N. Cale and G. Willoughby. *Nature*, London, 1948, Vol. 162, p. 411.
6. A. H. Sully. *J. Inst. Met.*, Vol. 80, 1951–1952, p. 173.
7. A. Preece and others. *Iron and Steel Institute Special Report No.* 24, 1939, p. 9.
8. W. C. Leslie and M. G. Fontana. *Trans. Amer. Soc. Metals*, 1949, Vol. 41, p. 1213.
9. J. L. Meijering and G. W. Rathernau. *Nature*, 1950, Vol. 165, p. 240.

FOUR

The Application of High-temperature Materials

In previous chapters the properties called for in alloys for service in various parts of the gas turbine have been discussed and some of the metallurgical factors influencing the development of these properties outlined. In this chapter a general survey of the types of alloy used for the main parts of the turbine, i.e. compressor, combustion chamber, turbine discs and rotor assembly, turbine blading and heat exchanger is given, followed by an outline of the composition and properties of the best-known commercial alloys. It must be stressed that, in the case of many of the low-alloy steels, while a certain composition may be well known under a trade name it may also be available from other manufacturers under a different name.

In reviewing the alloys used in a modern gas turbine, it must be remembered that, in many cases, the service conditions of a component will vary widely according to the power rating and type of gas turbine, i.e. aircraft, marine or industrial. Consequently it is not possible to do other than discuss the main types of alloys available, with an indication of their special properties and limitations. It must be borne in mind that, in addition to mechanical properties, there are many other factors which will govern the selection of a suitable alloy, such as corrosion and oxidation resistance, weldability, forgeability, density, etc. As may be seen from published details of alloys used in modern gas turbines, a designer frequently uses a stronger, though more expensive, alloy than may be strictly necessary in certain applications, when the greater reliability thus obtained more than offsets the higher cost.

Compressor

For the compressor blading the main requirements are:

(a) Good resistance to creep at temperatures up to 200–300°C (392–572°F) since the temperature of the air may rise to this level during compression.
(b) Good fatigue resistance.
(c) Ease of fabrication.
(d) In cases where the overall weight of the unit is important, such as the aircraft gas turbine compressor, the specific gravity of the material should be low.

(e) For marine purposes, and certain land-based turbines, the compressor blading should have good corrosion-resisting properties.

In the aircraft field the wrought aluminium alloys and, to a lesser extent, cast magnesium alloys are proving extremely satisfactory. Their low density keeps the overall weight of the compressor down, and their ease of fabrication when compared with steel allows the entire rotating section of the compressor to be forged or even cast as a single unit, where the design makes this possible. The creep strength of alloys based on these two metals have been much improved in the last five years and they are now capable of short-time operation up to the 250–350°C (480–660°F) required: "short-time" in such cases usually refers to periods of operation from a few hundred to 1000 or 2000 hours. Where lives of several years are called for, these alloys are not suitable.

The early aluminium alloys specially developed for operating at elevated temperatures were for service as piston material in internal-combustion engines. The first of these was known as Y-alloy and contained approximately 4% copper, 1·5% magnesium, 0·7% silicon, 2% nickel, 0·6% iron and 0·2% titanium. This composition is heat-treatable and possesses moderately high strength at temperatures up to 300°C (572°F) but has rather a low resistance to shock. Since this alloy was discovered, however, there have been many improvements. One such alloy, which was used for the impellers of the early gas turbines, is known as Hiduminium R.R.59. This alloy contains 2·2% copper, 1·5% magnesium, 0·9% silicon, 1·0% iron and 1·2% nickel, and is suitable for use up to about 200°C (400°F). A more recent alloy, R.R.58, is suitable for a maximum service temperature of 250°C (480°F), and is probably the best in this series of alloys. It contains 2·7% copper, 1·5% magnesium, 1·0% magnesium, 1·0% iron and 1·2% nickel. R.R.57 has also been developed for the range 250–300°C (480–572°F), and contains 6% copper and 0·25% manganese. Its properties remain reasonably constant even after holding for several hundred hours at temperatures up to 300°C (572°F). All these alloys have a high resistance to corrosion, particularly of the intercrystalline type, to which the thin trailing edges of the blades are most susceptible. The natural corrosion resistance of an aluminium alloy is greatly enhanced by an electro-chemical treatment, generally referred to as "anodising", which produces a highly-protective film of aluminium oxide on the surface.

The aluminium alloys are readily forged, particularly in the annealed condition and require little or no surface cleaning after-

wards. The cast aluminium alloys have so far had little or no application in the gas turbine in situations where both temperature and stress will be encountered, but they are frequently used for casings, etc., in aircraft gas turbines. They lose their mechanical properties rapidly when heated to 150–200°C (300–400°F) for prolonged periods.

Magnesium alloys are relative newcomers to the metallurgical field. The very high inflammability of magnesium and the difficult production of the pure metal delayed its use until recent years and, consequently, the application of magnesium alloys to the gas turbine is rather limited at the present time. It is possible, however, that as the reliability of magnesium alloys and their potential usefulness become more widely appreciated, the use of the alloys will become more widespread. At the present time a limited number of magnesium alloys are available for operation at temperatures up to 350°C (660°F) under the stresses encountered in compressor operation.

Magnesium metal crystallises in the hexagonal structure and consequently it possesses only a small capacity for cold working. However, both cast and wrought metals are used in practice; the former being preferable in creep-resisting applications, the latter having better mechanical properties at room temperature. The main alloying elements are aluminium, zinc and manganese. The first two increase the mechanical strength while manganese improves the resistance to corrosion.

The discovery that small additions of zirconium greatly improve the workability and mechanical properties of magnesium-zinc alloys, while decreasing the susceptibility to stress corrosion, has led to an increased interest in these alloys. The main compositions concerned are the 4% zinc -0·7% zirconium (Elektron Z.52 and R.25) and 3% zinc -0·7% zirconium -3% cerium alloys (Elektron Z.R.E.1 and M.C.Z.). These alloys, after suitable heat treatment, possess excellent creep and fatigue resistance up to 250°C (490°F), combined with good founding characteristics. One of the most recent of these alloys contains 3% thorium in place of cerium (Elektron Z.T.1 of the Magnesium Elektron Co.) and the improved creep properties make the alloy particularly suitable for use in the compressor. The heat treatment of these magnesium alloys must be carried out in a forced draught of air containing additions of sulphur dioxide and usually consists of a simple stabilising anneal at the maximum service temperature.

The specific gravities of aluminium alloys lie in the range 2·6 to 2·9, while for magnesium alloys the value is usually about 1·8, so that both types of materials are attractive where weight considerations are important. The low density of magnesium also

implies that its alloys would be subjected to appreciably lower creep stresses due to centrifugal forces than the aluminium alloys. In addition, the slightly higher maximum temperature at which their creep properties are maintained make them suitable for the hotter parts of the compressor. Unfortunately the fire risk with magnesium alloys is appreciable and limits their application at such temperatures. The hexagonal structure of magnesium alloys also implies a high inherent damping capacity, which is particularly important in the case of compressor blades, since the fluctuating stresses are particularly severe.

The recently developed titanium alloys are now appearing in commercial production, though the total annual world output is still only a few thousand tons. A suitable method for the large-scale extraction of this metal was only developed about 15 years ago by Kroll. It consists of the initial formation of titanium chloride and its subsequent reduction by magnesium. In this country, Imperial Chemical Industries Ltd. now have a plant producing about 1500 tons annually by a modified Kroll process using sodium in place of magnesium.

The considerable interest which has arisen in the application of titanium alloys springs from a number of very desirable properties, particularly the possibility of tensile strengths in the region of 85 ton/in^2 with only half the specific gravity of steel. Because of the rather rapid loss of strength and susceptibility to excessive oxidation as the temperature increases, however, the useful limit to titanium alloys is probably only in the region of 500°C (932°F). The properties of titanium alloys, therefore, are roughly intermediate between those of aluminium or magnesium and the ferritic steels.

Although titanium is still rather expensive, its increasing use in the aircraft field both for structural components and engine parts should eventually bring the price down to a more satisfactory level. In the aircraft gas turbine it has been successfully used for the compressor disc and blading as well as engine cowlings and exhaust shrouds. The alloys are undoubtedly particularly useful in the compressor since their properties are so much more suitable for the operating conditions encountered there than those of any other alloys. The extensive use of titanium in an aircraft gas turbine compressor can effect a saving in weight of approximately 20% of the total engine weight, a very real advantage.

I.C.I. Ltd. are at present supplying, in addition to two grades of titanium metal, four alloys of different strength levels. A long-term research programme is being embarked upon and the next few years will undoubtedly see a great increase in the types of alloys available. The alloy developed for its high-temperature creep

strength, titanium 371, is a ternary titanium-tin-aluminium alloy and undergoes less than 0·1% total strain under a stress of 30 ton/in^2 after 300 hours at 400°C (752°F).

In certain conditions the resistance of titanium to corrosion is higher than that of stainless steel, particularly to that caused by sea-water. Its application in the compressor of marine gas turbines should, therefore, prove particularly attractive, since the moisture-laden sea air is detrimental to many metals.

In its fatigue properties titanium is similar to steel in that it possesses a definite fatigue limit, i.e. the application of reversed stresses below about 50% of the ultimate tensile strength is unlikely to lead to failure.

In marine and land-based gas turbines, weight is not such an important consideration as in the aircraft engine and long-term reliability, particularly resistance to fatigue, becomes a prime factor. Furthermore the higher pressure ratios available from an axial compressor make this design more attractive than the centrifugal type. Consequently, these engines generally use the heavier but much stronger alloys such as aluminium bronze, copper-nickel alloys, low-alloy steel and stainless iron or steel.

Substantial numbers of both moving and stationary blades are manufactured from aluminium bronze, a copper alloy containing from 5% to 11% aluminium, together with iron, nickel and manganese in varying proportions. This material is readily formed into blade shape by precision forging and the surface finish is so good after this operation that little further treatment is required. The specific gravity is approximately 7·5. An alloy containing 10% aluminium, 5% iron and 5% nickel (e.g. Hidurax 1/12A) has particularly good creep strength. At the temperature encountered in the compressor, the thin surface film of alumina formed on these alloys provides complete resistance to oxidation and a high degree of corrosion resistance.

In certain gas turbines where a very high compression ratio is required a steel may be necessary for the compressor blading. Several types of steel are used, ranging from mild steel to the stainless materials. The greater density is offset by the very much higher resistance to fatigue stresses.

The corrosion resistance of stainless iron and steel, however, is frequently required and these are the most popular ferrous alloys for compressor blading. Stainless iron blading has been used for a number of years in steam turbine manufacture, the composition being about 12% chromium with the carbon content being kept as low as possible. Ferritic steels of the 12% chromium type and austenitic steels of the stabilised 18% chromium-8% nickel type are also widely used.

The manufacture of blading in these materials is conveniently carried out by precision casting or precision forging. A mechanical or electrolytic polishing process is usually employed as a final operation to ensure freedom from surface flaws, the presence of which would impair the fatigue properties of the blade.

All the above materials are "conventional" in the sense that they are produced by ordinary melting and casting techniques. A material developed in America and used in fairly large amounts for compressor blades lies outside this category and in the field of powder metallurgy. It comprises a skeleton of sintered iron powder, impregnated with copper and heat-treated. It is claimed that this process produces a high-strength blade more cheaply than the usual methods. An unusual material originating in Switzerland under the initials S.A.P., consisting of sintered aluminium powder and containing a small proportion of aluminium oxide, has also been used for the manufacture of compressor blades. In addition to being somewhat stronger than aluminium at ordinary temperatures, its mechanical properties are much less affected by increased temperature. It is manufactured in this country under the name Hiduminium 100 by High Duty Alloys Ltd. under the patents of the Société Anonyme pour l'Industries de l'Aluminium.

Combustion Chamber and Auxiliary Ducting

The materials chosen for the manufacture of the combustion chamber are subjected to the highest temperatures prevailing in the gas turbine. Consequently a major requirement of these alloys is good oxidation resistance. In addition the construction requirements call for good deep-drawing, bending and welding properties. Since vibrational stresses will be transmitted to the walls of the combustion chamber through the air stream from the compressor, the alloy must also possess a satisfactory fatigue strength.

To offset the high temperature of the burning gases, the designer usually arranges for a supply of secondary air to be diverted to flow through a series of slots cut in the wall of the flame tube. This air serves as insulation between the hot gases and the metal but even so the temperature of certain areas may rise to as high as 800°C (1500°F) since this method of cooling is rarely completely effective. The steep temperature gradients possible in these circumstances set up thermal stresses, which, unless the material is sufficiently strong, may cause warping and severe distortion. Good thermal conductivity will aid in reducing the degree of local overheating.

The necessary strength can be obtained by selecting a suitable thickness of high-strength alloy but since the construction of a deep-

drawn flame tube calls for a considerable amount of deformation in the course of manufacture, the greater the thickness the more difficult fabrication will be.

The most frequently used British material in this section of the gas turbine is probably Nimonic 75. This is a nickel-chromium alloy, stiffened by additions of titanium and aluminium, and one of the first of the alloys which were developed specially for the gas turbine. It is readily produced in sheet form, and can be bent or deep drawn to the required shape. It also possesses excellent welding properties.

Other alloys in the Nimonic range used in flame tubes are Nimonic C, Nimonic D and Nimonic F. Nimonic C is of similar composition to Nimonic 80 but is not supplied to specified creep data as is Nimonic 80. Nimonic D contains a large amount of iron: its composition is nickel 37%, chromium 18%, silicon 2%, with the balance iron, and is consequently not as strong nor as oxidation resistant as the higher nickel alloys but is capable of operation in oxidising atmospheres at temperatures up to 950°C (1750°F) under moderate stresses. Nimonic F is a modified form of Nimonic 75, the principal difference being a higher iron content, from 5 to 11% against the 5% maximum for Nimonic 75. In mechanical properties the two materials are very similar.

The latest addition to the Nimonic sheet alloys is a clad material comprising a copper sheet faced on both sides with Nimonic 75, called Nimoply. The copper improves the thermal conductivity of the Nimonic, so reducing the occurrence of local "hot spots".

Other alloys frequently used in the construction of combustion chambers include:

(a) H.R. Crown Max, a 12% nickel, 23% chromium austenitic steel stiffened by 3% tungsten and 1·5% silicon.

(b) Inconel, a nickel base alloy containing 14% chromium and 6% iron.

(c) Red Fox 31 and 33, both high-alloy steels, the former having 25% chromium and 15% nickel and the latter 20% chromium, 30% nickel and 1–2% titanium.

(d) Immaculate 5, an iron-base alloy having additions of 25% chromium, 21% nickel and 1·5% silicon.

(e) A number of stainless steels, variations of the well-known 18% chromium-8% nickel type of steel, having higher nickel contents with, in some cases, slightly more chromium as well. The higher nickel content completely suppresses the austenite to ferrite transformation, which is essential in view of the stresses which could be set up if transformation occurred in service. Hardening elements such as columbium,

molybdenum and titanium are added to increase the creep strength of the alloys.

(f) More highly alloyed steels and special alloys such as the several variations of 25% chromium-20% nickel austenitic steels, and the American alloys, Discaloy, N-155 (low carbon), Vitallium, etc.

Typical analyses of all the alloys mentioned above are given in the Appendix of Compositions (*see pages* 153–158).

The tubes through which the fuel is injected into the air stream are frequently made from alloys such as Nimonic 75, Nimonic F, Inconel, etc. In this position the main requirements are oxidation resistance, and with certain fuels, stability in reducing conditions to sulphurous gases.

In the U.S.A. a technique involving the coating of the interior surfaces of the combustion chamber with a refractory substance was successfully developed from wartime research aimed at increasing the life of turbo-supercharger exhaust cones. The difficulties involved in the production of satisfactory refractory coatings are dealt with in a later chapter.

Certain industrial gas turbines now in production include a refractory brick lining for the combustion chamber. The B.T.H. Company's 2500-kW units for the East African Power and Lighting Co., for example, have a 1½-in. thick refractory lining of 80% alumina material in the combustion zone. Experience with the gas turbine installed in the Shell tanker *Auris* has shown that the refractory brick lining to the combustion chambers only needs replacing about every 15,000 hours.

The introduction of an insulating layer between the flame tubes and the outer combustion chamber casing is becoming more frequently used. A number of mineral wool blanketings are available, a typical one being made by British Refrasil Co. Ltd. Refrasil blankets consist of a vitreous silica fibre, which is suitable for use at temperatures up to about 1093°C (2000°F), enclosed between thin walls of metallic foil, generally 18/8 stainless steel or Inconel. Other companies which have manufactured similar products include Delaney Gallay Ltd. and Burnley Aircraft Products Ltd. The use of these blankets greatly lowers the temperature to which the combustion chamber casing is subjected.

Discs and Rotors

The disc and rotor of the gas turbine are subjected to a wide variation of stress and temperature across the section and consequently a very high-quality alloy is required.

The principal stresses set up in these components arise from the centrifugal forces and the steep thermal gradients. The value of the centrifugal stresses depends on the speed of rotation, the size of the component and the density of the alloy. These mechanical stresses, which increase as the square of the speed, may give rise to gradual stretching of the material or even "bursting", if a sufficiently high speed is attained. Consequently a high proof stress is a necessary property of the selected alloy.

Since the rim temperature is high, of the order of 550–600°C (1000–1100°F) due to direct exposure to the hot gases emerging from the combustion chamber, while the hub temperature is only of the order of 300°C (572°F), steep thermal gradients are set up. The values of the thermal stresses arising under such conditions are governed by the temperature distribution, the coefficient of thermal expansion and the Young's Modulus of the material.

The main requirements of rotor and disc alloys may be summarised therefore as:

1. High proof stress.
2. Good creep strength.
3. Oxidation resistance.
4. Good thermal conductivity.
5. Low coefficient of thermal expansion.

The early gas turbines used austenitic 18% chromium-8% nickel alloys, suitably modified, and the rotor and discs were forged as a single unit. It was later realised, however, that by efficient cooling the maximum temperature in the rotor could be kept below 550°C (1000°F), thus permitting the use, in certain cases, of ferritic alloys. At the present time, therefore, both types of alloy are in use, according to the type of turbine concerned.

The relative advantages and disadvantages of ferritic and austenitic alloys are summarised in Table VI.

The construction of the rotor assembly for the modern gas turbine presents a formidable problem. For the larger sizes it is extremely difficult, first, to produce the large sound ingot required and, secondly, to forge the ingot to the required shape. The successful production of built-up rotors, on the other hand, depends mainly on the development of good weld metals and their efficient deposition. The relatively unknown creep properties of weld deposits and the necessity for the after-treatment of welded constructions have tended to restrict the manufacture of composite rotors, but there are several in operation, apparently with success.

In America, all the discs manufactured by one large producer (General Electric Co.) are of a composite construction, an austenitic

TABLE VI

The advantages and disadvantages of ferritic and austenitic alloys

FERRITIC	
ADVANTAGES	DISADVANTAGES
1. Good ductility. 2. High proof and maximum stresses, at lower temperatures. 3. Lower coefficient of thermal expansion. 4. Higher thermal conductivity. 5. Good forgeability. 6. Good machining properties. 7. Structural quality, especially in the central portion of the ingot. 8. Lower cost, about one-quarter to one-half that of a similar austenitic disc. 9. Lower strategic metal content.	1. More susceptible to loss of strength if design temperature is exceeded. 2. Poor weldability, and complete heat-treatment required after welding. 3. Poor oxidation resistance. 4. More liable to segregation during casting.
AUSTENITIC	
ADVANTAGES	DISADVANTAGES
1. Superior hot-strength properties. 2. Better oxidation resistance. 3. More easily welded and need only be stress-relieved after welding.	1. Require a large amount of strategic elements. 2. More difficult to forge. 3. Greater cost, due to 1 and 2. 4. Tendency to solidify in ingot with central unsoundness.

alloy rim "16-25-6" being welded to a ferritic hub (a Ni-Cr-Mo steel), and this combination appears to be perfectly satisfactory in service. In the case of a complete rotor assembly, it is probable that the use of an austenitic forging for the hotter end and a ferritic forging in the cooler stages would substantially reduce the initial cost if the different properties could be reconciled.

The stretching and bursting tendencies of discs when running at high speed have already been mentioned and the need for a high proof stress emphasised. Research has shown that a considerable improvement in the tensile and creep properties may be brought

about by a small amount of deformation carried out at 650–900°C (1200–1650°F), i.e. below the normal forging temperature range. This technique is now frequently applied to certain rotor and disc alloys and is known as "warm-working".

Table VII indicates the alloys most frequently used in this part of the gas turbine at the present time. Other alloys have been developed and used in specific instances but those listed account for the majority of the rotors and discs in use at the present time.

TABLE VII

Materials used for turbine discs and rotors

FERRITIC ALLOYS
Molybdenum-vanadium steels—e.g. Mova.
1% chrome-molybdenum steels—e.g. Hecla. 153.
1% chrome-molybdenum-vanadium steels—e.g. H.51.
3% chrome-molybdenum-vanadium steels—EN.40 type.
3% chrome-molybdenum-vanadium-tungsten steels—e.g. ChromVa-W.
12% chromium steels—e.g. F.G., H.G.T.4, H.46, Silver Fox 11, etc.
AUSTENITIC ALLOYS
18% chromium-8% nickel type—e.g. F.C.B. (T).
Nickel-chromium steels—e.g. R.20, G.56.
Nickel-chromium-cobalt steels—e.g. G.18B, Rex 326.
Cobalt-nickel-chromium alloys—e.g. G.32.
Nickel-chromium alloys—e.g. Nimonic 80, 80A and 90.

Ferritic Steels

The low-alloy steels are restricted to operating temperatures below about 575°C (1065°F) in the case of the aircraft gas turbine and below 520°C (968°F) where much longer life is called for. Above these limits the creep properties are rapidly deteriorating. Where suitable cooling arrangements can be provided it is possible to use gas temperatures in excess of these levels while still restricting the steels to a suitably low temperature. The increasing alloy content of the steels provides for a range of tensile properties according to the design requirements. There is no significant difference in creep properties in the steels in this category, with the exception of the 12% chromium steels which permit the use of temperatures up to 550–570°C (1022–1058°F) for prolonged periods.

Austenitic Alloys

The steels developed from the basic 18% nickel-8% chromium steel contain a variety of stiffening elements, such as molybdenum,

vanadium, columbium, tungsten, etc. They have useful creep properties up to about 650°C (1202°F) and good hot-working properties in large sizes, a valuable characteristic for the manufacture of large rotors.

For better creep properties and service temperatures above 650°C (1202°F) there are a variety of alloys containing nickel, chromium, cobalt and iron in different proportions. The number of alloys containing large amounts of cobalt is not very great since this is an extremely costly metal and also it has a strategic value which imposes a limit on the extent to which it can be used.

The Nimonic alloys, Nimonic 80, 80A and 90, have been developed from an 80% nickel-20% chromium alloy by the use of strengthening additions, chiefly aluminium and titanium. They possess excellent creep properties coupled with good resistance to fatigue and oxidation. Nimonic 90 contains a proportion of cobalt (15–21%) in place of some of the nickel. Although there are instances of these alloys being used for rotor discs in certain land and marine gas turbines, their very high cost and their more suitable use as turbine blade materials at the higher temperatures and stresses for which they were developed have limited their use in this section of the gas turbine.

American Alloys

Of the many American "super-alloys" developed for high-temperature service, 19-9DL and 16-25-6 are the principal rotor and disc alloys. The former is an improved 18% chromium-8% nickel steel, using titanium, tungsten and columbium as strengthening elements. The latter contains 16% chromium, 25% nickel and 6% molybdenum, from which the name is derived, and is interesting in that it contains more nitrogen (0·15%) than carbon (0·10%).

From the wide range of alloys available it can be seen that the selection of a suitable material is largely governed by the operating temperature and the stresses likely to be set up. The present trend, particularly in the aircraft gas turbine, is towards the provision of an effective flow of cooling air at the face of the disc, thus permitting the use of gas temperatures of 700–800°C (1300–1475°F).

Turbine Blading

The limiting factor in the use of higher operating temperatures in the gas turbine has always been the development of materials for the rotating blades of the turbines. Several new alloys have been introduced as a result of the large-scale creep testing carried out by the alloy producers in recent years, but there are now only small increases in maximum operating temperatures to be obtained and

the accent is mainly on increasing the creep strengths of the existing alloys at the present temperature levels. Any significant increase in this temperature range will probably come from the successful production of a non-metallic or metal-ceramic material.

In the following table the materials now in use as blading alloys are given:

TABLE VIII

GREAT BRITAIN
Low-alloy steels—coolest stages of certain industrial turbines only. Medium-alloy steels—e.g. F.C.B. (T), Rex 326, G.40. Nickel-chromium-cobalt alloys—e.g. Hecla EM.20 and MM.20, G.32 and G.42B. Nickel-chromium alloys—e.g. Nimonic 80, 80A, 90, 95 and 100.
AMERICA
Medium-alloy steels—e.g. 19-9DL, Discaloy 24. Nickel-chromium-cobalt alloys—e.g. N-155, S.590, S.816. Nickel-chromium alloys—e.g. K.42B, Inconel X, Hastelloy B and C. Cobalt-based alloys (e.g. Vitallium or Haynes Stellite 21, 422–19, 6059 and X. 40—all used in the cast condition.)

It has always been the practice in this country to use only forged rotor blades, whereas American manufacturers have, until fairly recently, favoured precision cast blades. The British preference for forged blades arises from the much greater uniformity in the properties of blades produced in this way. The cast cobalt alloys, however, are among the most creep-resistant alloys available and it is generally found that American gas turbines using these alloys run at higher temperatures than usual. Several alloys are available in this country for producing cast turbine rotor blades, e.g. G.34, Hecla EM.35 and MM.35, but little use has been made of them so far in this position. The present trend towards the use of shroud rings on the rotor blades will, however, encourage the use of these alloys, since the greater damping-out of vibrations that this will achieve will make the lower fatigue strength of cast blades of less importance.

The Nimonic alloys have so far provided the vast majority of British aircraft turbine rotor blades. Those which have found the most extensive use to date are Nimonic 80, 80A and 90. Nimonic 95 and 100 are fairly recent developments and are mainly designed for short-life requirements at fairly high temperatures.

While the nickel alloys, are, without doubt, superior to all the

iron-base alloys at the highest operating temperatures, the much greater initial cost of nickel alloy blades frequently restricts their use in land and marine turbines to the hottest stages. This may be illustrated by reference to the recent 2500-kW power sets built by the B.T.H. Co. Ltd. for the East African Power and Lighting Co. Here there are nine stages of rotor and stator blades, the first stage being Nimonic 80, the second Rex 326 and the rest F.C.B.(T), for both types of blade.

It was mentioned above that the cast cobalt alloys have greater creep resistance than any of the heat-resisting steels or other wrought heat-resisting alloys. The metallurgical structure of these alloys is similar to that of a cored solid solution with the minor phases present in the interdendritic region. It is interesting to note that if such a structure is annealed at a very high temperature so that homogenisation takes place, the alloy will lose a large proportion of its creep strength. Consequently the cast alloys are usually put into service with no prior heat-treatment.

Nozzle Guide Vanes

The main requirement in the alloy selected for the turbine stator blades, or nozzle guide vanes as they are sometimes called, is a high resistance to oxidation. The blades are usually forged and machined to shape, although some manufacturers use precision casting techniques, since the service stresses are relatively low, of the order of 1–2 ton/in^2.

The alloys used in this country range from the simpler stainless steels in the cooler stages, through the higher alloy steels, e.g. 25% chromium-20% nickel, in intermediate conditions, to the Nimonic alloys in the hottest region. In American turbines alloys such as S.590, N-155, Hastelloy C, Inconel X and Haynes Stellite 21 are frequently used.

A certain amount of research work has been carried out, and is still in progress, to develop nozzle guide vanes from either wholly refractory materials or metal-ceramic combinations. For example, Pamatrada are experimenting with zircon plus 2½% chromic oxide and silicon carbide blades, while in America Kentanium is promising. These materials are dealt with in Chapter Five.

Heat-exchangers

Under this heading the heat-exchangers of both the recuperative and the regenerative types used with the open-cycle turbine, the air heaters required in the operation of a closed-cycle gas turbine and the water heaters used in certain industrial turbine circuits to utilise the hot waste gases for the production of steam are considered, since the properties required are similar throughout.

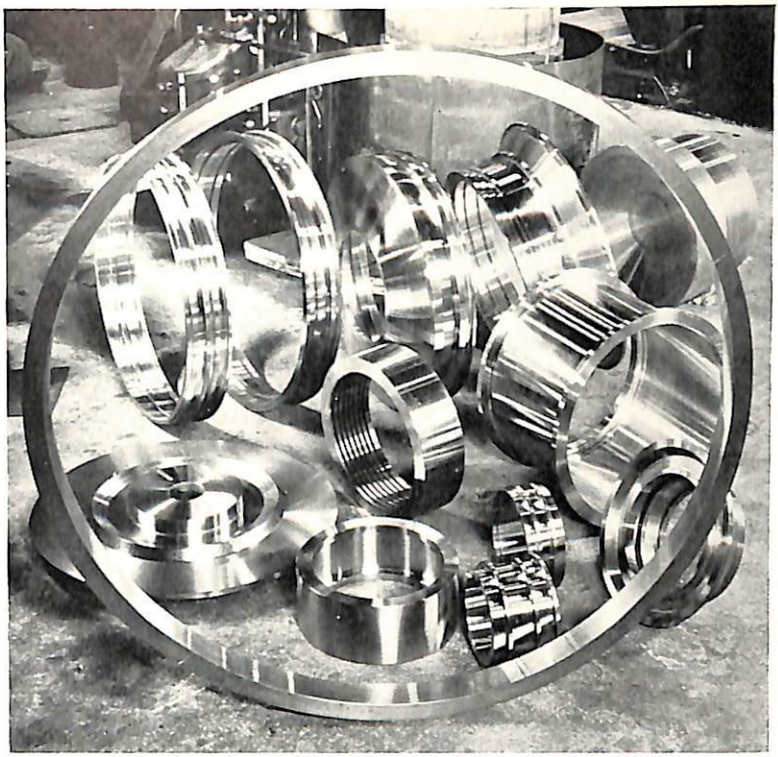
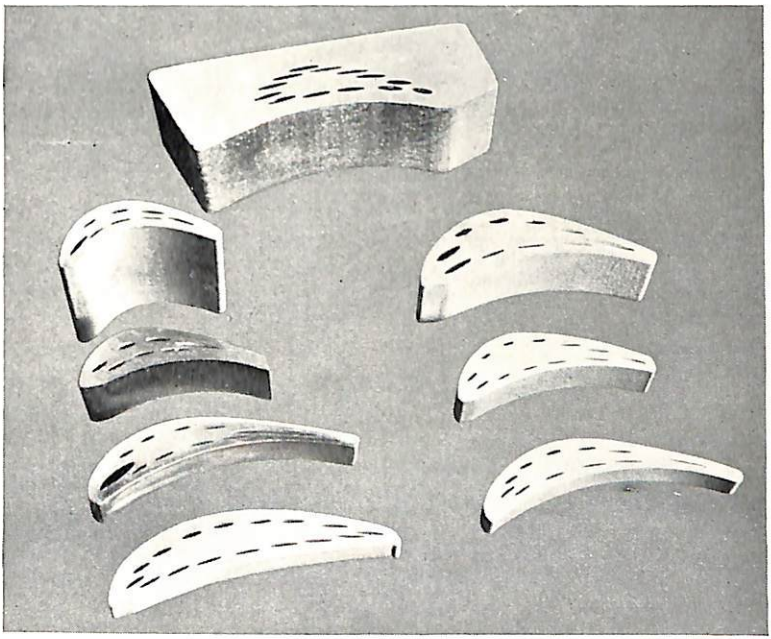


Plate 13. *Centrispun components, as manufactured by Firth Vickers Stainless Steels Ltd*

Plate 14. *Cooled rotor and stator blade sections produced by the extrusion of drilled and plugged Nimonic* 90 *billets*



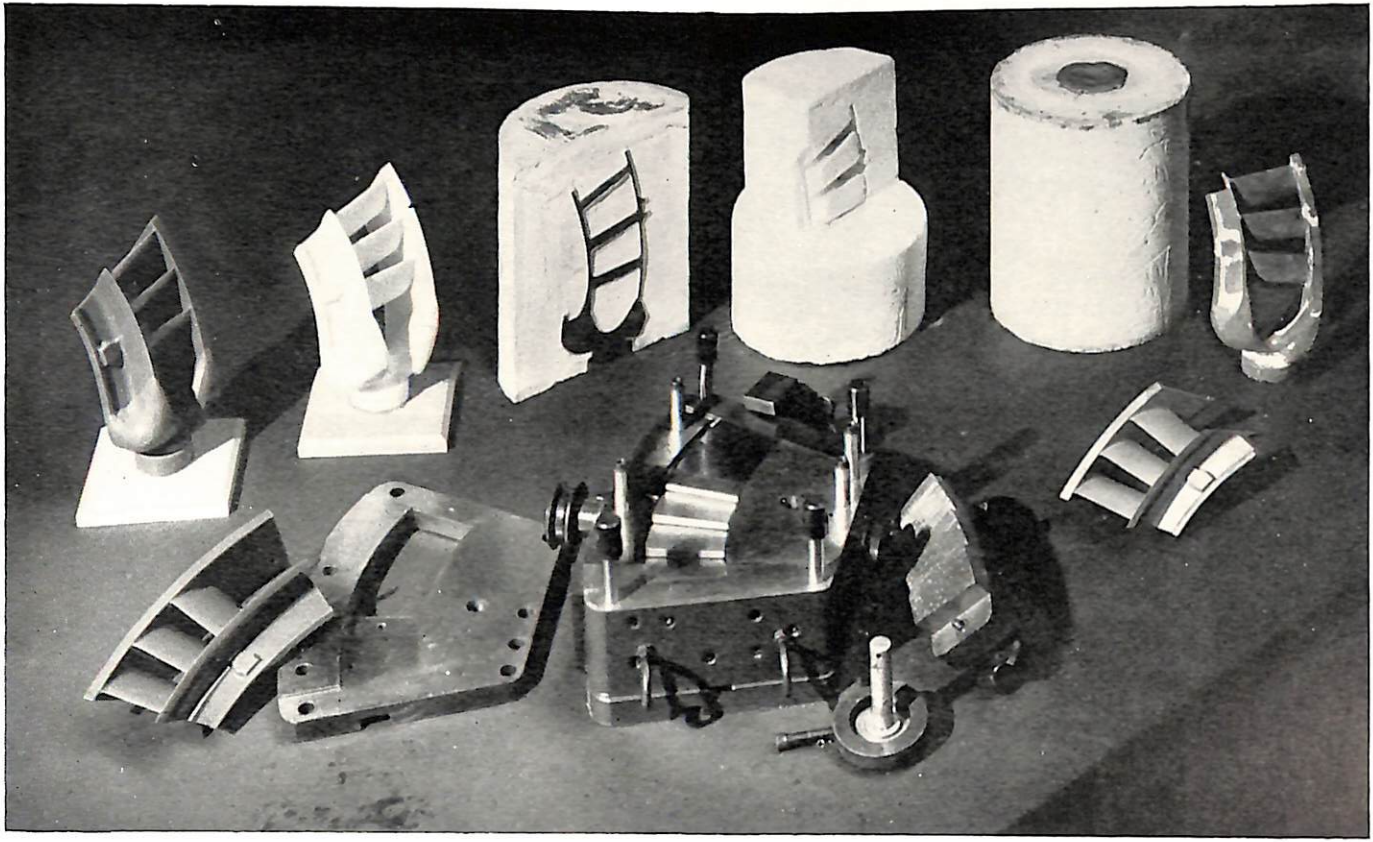


Plate 15. *The lost wax process for the production of cast turbine nozzle blades as used in Bristol gas turbines*

The most frequently used heat-exchangers are of the recuperative type, i.e. the hot gases flow over tubes carrying the air to be heated. The process is continuous.

The tubular heat-exchangers present several metallurgical problems which have to be solved with due regard to other factors such as cost, space, etc. The equation for the heat flow in such a case is:

$$q = \frac{kA(t_1 - t_2)}{D}$$

where q = total heat transmitted, in B.t.u./h,
D = wall thickness of tube, in feet,
t_1, t_2 = temperature of the hotter and colder sides of the tube respectively, in °F,
k = thermal conductivity of the tube material in B.t.u. ft/ft² h °F
and A = area of surface, in square feet normal to the direction of heat flow.

Consequently, it follows that for the same temperature conditions the maximum rate of heat transfer for a given material is obtained with a large surface area and a thin separating wall. Also, since air is not a particularly good conductor of heat, it is essential to arrange for turbulent flow inside the tubes, so that the air in the immediate vicinity of the hot wall is being constantly changed. These conditions are satisfied by the use of a very large number of small-diameter, thin-walled tubes, set closely together in a gas-tight shell. The latter is usually a sheet-metal structure, suitably welded.

Turning therefore to the factors governing the selection of a heat-exchanger alloy it is clear that the material must possess the following properties:

(a) High thermal conductivity.
(b) Good oxidation and corrosion resistance.
(c) Formability, for drawing into tubing.
(d) Weldability, to make the joining of the many sections a simple operation.
(e) Low cost, since large quantities are required.
(f) Sufficient mechanical strength at the service temperature to prevent distortion and sagging.

Some typical thermal conductivity data for various metals are given in the table overleaf.

In general an increase in temperature leads to a decrease in the thermal conductivity in the case of a metallic material.

From the table it is clear that copper-base alloys are preferable, since the rate of heat transfer is so high that a smaller heat-exchanger

could be used for the same degree of recuperation. The addition of aluminium and nickel to copper provides the necessary oxidation resistance and an appreciable resistance to corrosion. The strength of these alloys is a limiting factor in service and at present these alloys are restricted to about 500°C (932°F). The maximum temperature required in a gas turbine heat-exchanger, however, may be as high as 800°C (1475°F) (in the case of an air heater for a closed-cycle unit) and, for these higher ranges, a more heat-resistant alloy is required. According to the conditions, suitable alloys are to be found based on the 25% chromium-20% nickel, or the 18% chromium-8% nickel steels. For higher resistance to corrosion there are several nickel-base alloys, such as Inconel, and the Nimonic alloys 75, C, D and F. With certain nickel-containing alloys, however, care must be taken to see that they are not subjected to high concentrations of sulphur-bearing gases unless the atmosphere is highly oxidising. Under reducing conditions these alloys are liable to severe corrosion, sometimes known as "green rot".

TABLE IX

Thermal conductivity of some metals at 500°C (932°F)

MATERIAL	THERMAL CONDUCTIVITY	
	cal cm/cm² sec °C	B.t.u. ft/ft² h °F
Copper	0·86	20·8
Aluminium ..	0·45	10·9
Nickel	0·130	31·5
Iron	0·106	25·6
Mild steel ..	0·09	21·8

Monel, a copper-nickel alloy containing 67% nickel, 30% copper and small amounts of iron and manganese, is also used for heat-exchanger tubing. K-Monel is an age-hardening form of this type of alloy and is consequently stronger.

Low-alloy steels, such as molybdenum steel, are occasionally used in the lower temperature range, but are unsuitable for service where corrosion is liable to be troublesome.

For moderate working temperatures aluminium alloy tubes may be used, though the rapid loss of strength at temperatures above 200–300°C (400–500°F) limits the use of these materials to conditions of low stress.

Although it is not yet common practice, it is possible to use the

more expensive heat-resistant alloys at the hottest end of the heat-exchanger, changing to the low-alloy steels or aluminium bronze at intermediate temperatures, and possibly to aluminium alloy at the cooler end. In such cases, however, problems such as those of joining the different materials without introducing stress-corrosion conditions may arise. Since the coefficients of thermal expansion of these various materials are widely different, the joints would tend to be points of high stress.

The tubular type of heat-exchanger is rather cumbersome when compared with the size of the gas turbine itself and several suggestions have been put forward for cheaper or smaller units based on similar principles or on the regenerative process. These may be briefly summarised:

RECUPERATIVE

(a) a variation of the tubular construction in which corrugated sheet material alternates with plain sheets, the apertures so formed acting as gas and air channels. In addition to reducing construction costs and utilising the available space to the fullest advantage this type of heat-exchanger permits the use of alloys which have only a limited formability and are difficult to produce in tube form.

(b) to improve the rate of heat transfer from the gas to the air, it is suggested that a liquid layer should be interposed between the two streams. The high thermal conductivity of the liquid would rapidly disperse the heat absorbed in the vicinity of the tube wall into the body of the liquid and, similarly, the heat absorbed by the cold air will be rapidly replaced. Against the adoption of such a system is the more complicated construction required and the possibility of corrosive conditions resulting in puncturing of the tubes. Since the thermal conductivity of the liquid would have to be very high, the most appropriate would be one of the low-melting-point metals or alloys.

It is fairly certain that this type of heat-exchanger will be used in conjunction with the nuclear power plants. The heat generated in the pile will have to be abstracted either by a gas or liquid metal. In either case the working fluid used will have to be reasonably immune to the intense radiation within the pile and so will the materials used to transmit it through the pile. If a certain amount of radio-active material is formed in the working fluid in passing through the pile, the decay of the radio-activity as it passes through the heat-exchanger may have some influence on the materials of construction. A good deal of research is going on into the effects

of radiation on metals and metal-gas reaction but few details of the work are published.

Regenerative

(a) The usual type of regenerative heat-exchanger involves the frequent reversal of gas flow through the two chambers and the use of large structures of refractory brick-work. It has recently been considered feasible to use a rotary type, in which the heat-absorbing media is divided into two sections and the direction of rotation normal to the gas flow. The gas and air conduits would be arranged so that while the hot waste gases were heating one half, the incoming air would be cooling the other half. The intermittent rotation of the regenerator would be automatic, based on measurements of the temperature of the gas stream after passing through the system. The main objection to this type of construction is the serious loss of efficiency caused by leakage, which is so difficult to prevent in a rotating system.

OTHER GAS TURBINE COMPONENTS

High-pressure Turbine Casings

The turbine casing needs to be sufficiently oxidation resistant to withstand the effects of the hot gases passing through it and of adequate strength to avoid warping troubles. The high-alloy steels, containing about 25% chromium and 12% nickel, or the modified 18/8 steels are generally satisfactory.

Bolts

Many parts of the turbine have to be inspected at regular intervals and it is therefore preferable to fasten the casings or cover plates in position with bolts. Since the pressure inside the turbine is of the order of several atmospheres, these bolts are under stress and if they do not perform satisfactorily the hot gases would tend to leak out at the joined faces instead of passing on energy to the turbine. Bolts are also used in certain turbines for fastening the discs to the main rotor shaft. For moderate temperature conditions the chromium-molybdenum, chromium-molybdenum-silicon and chromium-molybdenum-vanadium steels are generally used while the Nimonic alloys are suitable for use under more stringent conditions.

Shroud Rings, etc.

While the use of shroud rings on the moving blades of the turbine is usually undesirable from weight considerations, they are frequently used on the stator blades, to prevent warping and bending. The

materials are generally the same as those used in the blades themselves to avoid differential expansion effects. The method of construction of these ring-shaped components has generally been to centrifugally cast but there has recently been a trend towards extrusion. The Ugine-Sejournet process, using glass as a die lubricant, is favoured since it permits the extrusion of much stiffer alloys than the more common die lubricants, e.g. graphite. The extruded bar is subsequently bent to shape and welded.

Combustion Chamber Casing

In the industrial gas turbine this component is not subjected to very serious temperature or stress effects, particularly where a refractory lining has been used in the combustion chamber. Thus materials such as stainless iron could be used. In an aircraft gas turbine, however, light construction is adopted to reduce thermal stresses and give a minimum overall weight. Thus alloy-steel sheet or mild-steel sheet, the latter generally being nickel plated or aluminium sprayed to prevent corrosion, have been used. In such cases the use of several layers of insulating blanket in between the flame tube and the outer casing can greatly reduce the temperatures to which the outer casing might otherwise have been subjected.

Compressor Outer Casing

Again this component is not subjected to very high stress and temperature effects. Where overall weight is important, frequent use is made of aluminium or magnesium alloys. In other cases spheroidal graphite cast iron is a very useful material. The addition of magnesium to cast iron causes the graphite to be formed in spheroidal form instead of the large flakes obtained when no treatment is given. This leads to a doubling of the tensile strength and a considerable reduction in the brittleness.

If a high degree of corrosion resistance is required in the compressor casing, stainless iron or a chromium steel could be used.

The Diffuser

In the centrifugal compressor the air is picked up by the impeller blades and given a high velocity as it travels outwards. A number of curved vanes are then inserted in the path of this air to smooth out the flow and convert the high velocity into pressure. These vanes constitute the diffuser and they operate under the hottest conditions in the compressor. In an aircraft gas turbine they may be cast as an integral unit in magnesium alloy. For longer life and more severe conditions of service, however, steel diffuser vanes are preferred. Where corrosion resistance is required the steel may be of the stainless type or surface protected, e.g. by aluminising.

FIVE

Special High-temperature Materials

As a result of intensive research into the development of creep-resisting alloys, it is now generally acknowledged that, although many new alloys may be produced for the operating temperatures now in use, the ultimate for the simple metallic blade almost certainly does not greatly exceed 800–900°C (1475–1650°F) if the life of the blade is to be greater than a few hundred hours. Precipitation-hardened alloys capable of sustained service at this temperature level would be so stiff that the ductility at room temperature would be almost nil and the blade highly susceptible to shocks arising from mechanical or thermal causes.

As the limits to the use of metal blades have become clear, attention has been directed to non-metallic materials. Many compounds have been produced in this field in recent years with very good high temperature properties and it is certain that creep and fatigue values at least equal to, if not better than, those of the best heat-resisting alloys in use at the present time are possible. Compounds such as the carbides, nitrides, silicides and borides of metals such as tantalum, columbium, titanium, silicon, tungsten, chromium, molybdenum, etc., have very high melting points and retain their hardness and high strength to temperatures in the 1000–1200°C (1800–2200°F) range in many cases. Their complete lack of ductility, however, is a serious disadvantage, since the slightest shock results in fracture.

Consequently, on one hand there are the metals in which precipitated carbide (or nitride, etc.) particles give greatly increased stiffness, and, on the other, the sintered carbides themselves with a high degree of hot strength but negligible ductility. It was a logical development from this to produce an intermediate material by making small metal additions to the non-metallics in an endeavour to provide the required ductility.

Combinations of carbides and metal have been in use for many years in applications where high hardness and abrasion resistance are required, such as tips for cutting-tools and wire-drawing dies, etc. The basic constituent of the majority of these cemented carbide materials is tungsten carbide, since this is one of the hardest materials known. It is prepared by heating finely powdered tungsten in the presence of lamp-black at temperatures of the order of 1375–1400°C (about 2500°F). The resulting product is ground

and screened to various sizes before use. In addition, the carbides of titanium and tantalum are in use, these being prepared by heating the respective oxides with carbon at temperatures up to 1650°C (3000°F).

The experience obtained in the manufacture and use of these materials has been of great benefit to the development of similar carbide-metal combinations for the gas turbine.

Manufacturing Methods

The selected carbide or carbide mixture is generally milled with the metal binder for a prolonged period, during which time the particles become coated with the metal. This greatly facilitates the subsequent sintering operation but is not always necessary.

The mixture is then shaped by one of the following methods:

(a) *Cold Pressing*—this consists of pre-shaping the carbide-metal mixture in small presses, followed by a sintering treatment carried out below the melting point of the metal. This confers a sufficient degree of strength to the pressing to permit easier handling. The semi-sintered material is then given a final forming operation, e.g. machining, and fully sintered at a temperature above the melting point of the metal. The molten metal flows into the voids and, by surface tension effects, pulls the carbide particles firmly together. This method is usually restricted to simple shapes. In applying a high pressure to a powder, care has to be taken to avoid uneven pressure distribution throughout the body of the material, since this gives rise to zones of weakness. An improvement on the cold-pressing method, therefore, is that known as the "rubber-sleeve" process, in which a rubber mould is filled with the powder and subjected to hydraulic pressure. This provides an even pressure application over the surface of the powder and permits the production of irregular shaped pressings.

(b) *Hot Pressing*—this process combines the several operations outlined in (a) into one. By means of suitably designed presses and the use of graphite dies, the temperature is rapidly raised while the material is held under pressure. The total time required for the manufacture of the final shape is thus greatly reduced, but this method is rarely used in commercial production.

(c) *Extruding*—the operations are similar to those outlined in (a) but extruding presses are used to shape the material, suitably "plasticised" by addition of waxes, etc., in the early stages.

(d) *Slip-casting*—this method is useful for the production of simple hollow components. A metal-ceramic-water mixture is poured into porous moulds and the water drained away by capillary action through the walls. When sufficiently dry the powder, which forms a layer on the walls corresponding to the outline of the mould, is removed and fired.

(e) *Cold-casting* (a patented process)—this comparatively new technique consists of mixing the required powders with a special organic fluid which will polymerise at, or slightly above, room temperature in a very short time. The mixture is poured into wax moulds of any desired shape. When polymerisation is complete the wax can be melted away and the solidified material, after slow heating to drive off the organic compound, finally fired.

A different type of process consists of preparing a "skeleton" of a refractory material by pressing after mixing with an easily removable binder such as wax or sawdust. After carefully burning out the latter the refractory may be heated to a sufficiently high temperature to effect a partial sintering, which gives a good bonding between particles but leaves a certain porosity. If the skeleton structure is now brought into contact with liquid metal it acts as a sponge and the metal enters the voids. This "infiltration" process provides a ceramic-ceramic bond instead of ceramic-metal-ceramic. The final density of this type of product is never quite equal to the theoretical value since some voids tend to become isolated during the sintering operation and the metal cannot penetrate into them. The method is, however, extremely useful where the sintering temperature of the ceramic is considerably higher than the melting point of the metal.

Applications to High-temperature Materials

The very hard tips for cutting tools have been manufactured for many years by one or other of the methods described above. The particles of carbide have excellent cutting properties and the admixing of from 3% to 12% of cobalt provided the necessary shock resistance for this application. Cobalt combines with tungsten carbide so that during the heat-treatment the particles of carbide become coated with a thin layer of the compound and a very good bond is formed.

The addition of about 10% tantalum carbide is used to greatly increase the wear resistance, since this material is extremely resistant to abrasion, while the presence of either titanium or

columbium carbides in similar amounts confers a high degree of oxidation resistance. Titanium carbide mixed with about 20% cobalt forms a high-temperature material capable of use for prolonged periods up to 1000°C (1800°F), since titanium carbide is not readily oxidised. This material is however rather susceptible to shock and would be unsuitable where this is likely to be encountered. Very little data on the mechanical properties of materials of this type have been published but the following tensile strength values have been released after experiments carried out in the U.S.A. under a N.A.C.A. project:

Material	*Tensile strength*	*Temperature*
80% TiC+20% Co	25·8 to 45 ton/in²	870°C (1600°F)
86·7% TiC+13·3% Ni	7·2 ton/in²	1393°C (2530°F)

All these materials are characterised by their extreme hardness, excellent compressive strength and a high modulus of elasticity. The latter implies almost a complete absence of ductility and little measurable elongation is found in tensile tests. The strength in compression is much higher than that in tension, as is usual with brittle materials.

The high room-temperature hardness does not fall appreciably even at temperatures of 600–700°C (1100–1300°F) and any decrease on heating to higher temperatures is generally regained on cooling. This represents a considerable advantage over precipitation-hardened alloys which are greatly weakened by "over-ageing".

When titanium carbide is used, the solubility or combination with nickel is extremely limited, so that the particles are not as firmly bonded as is desirable. To combine the greater shock resistance of the tungsten carbide with the oxidation resistance of the titanium carbide it is possible to use mixtures of tungsten carbide with a tungsten-titanium double carbide (a patented process) and in this way the bonding effect is largely retained. For high-temperature use nickel is preferable to cobalt from the oxidation point of view, but this involves a slight decrease in the strength. This can be counteracted by alloying the nickel with a small proportion of tungsten.

A great deal of research has been carried out to determine the nature of the bonding involved in these composite materials. Several types of bond have been reported (1):

(a) Mutual solubility and compound formation, such as that outlined above for the tungsten carbide-cobalt and tungsten carbide-nickel materials.
(b) Mutual solubility with a surface phase, e.g. the bond between alumina particles and chromium is thought to be the result

of the formation of Cr_2O_3 on the surface of the metal and the subsequent solution of this oxide in the Al_2O_3 lattice.

(c) Reaction with a surface phase, e.g. in the case of alumina-iron mixtures the bond may arise from the formation of FeO and combination of this oxide with the alumina to produce the spinel $FeO.Al_2O_3$.

(d) Wetting of the non-metallic particles by the metal with no apparent solution or compound formation is frequently observed, the mechanical bond alone presumably being sufficient. The silver-graphite mixtures used as electrical contacts are typical of this group. This bond can sometimes be greatly improved by increasing the wetting power of the metal through the addition of suitable alloying elements.

Several other factors will govern the selection of the components of a non-metallic/metal combination for turbine blades:

1. The main stress set up in the blade arises from the centrifugal force, which depends on the density of the blade. Some idea of the considerable variation in the specific gravity available among the standard compositions is indicated in the table:

TABLE X

MATERIAL	SPECIFIC GRAVITY
Tungsten carbide..	15·67
,, ,, +3% cobalt	15·25
,, ,, +6% ,, ...	14·85
,, ,, +20% ,, ...	13·55
,, ,, +10% titanium carbide +6% cobalt ..	11·2
,, ,, +20% ,, ,, +7% ,, ..	9·0
Titanium carbide	4·91
,, ,, +20% cobalt	5·5

It will be noted that the use of titanium carbide is immediately attractive from weight considerations. The specific gravity of heat-resistant alloys usually lies in the range 8–12.

2. Good thermal conductivity is desirable, so that local overheating with its accompanying thermal stresses may be avoided. For similar reasons the thermal expansion of the metal and non-metallic material should be of the same order of magnitude, as otherwise the resulting combination would be subjected to serious internal stresses, which could result in distortion or even fracture.

3. Resistance to oxidation at high temperatures is required and

this is obtained by selecting the non-metallic components so that they are reasonably stable in hot oxidising conditions. Certain nitrides and borides, for example, are more resistant to oxidation than the majority of the carbides. In addition the oxidation resistance may frequently be improved by the production of a double carbide with chromium carbide or some other oxidation resistant carbide. The incorporation of chromium into the metallic matrix, while tending to weaken the bonding action, greatly improves the oxidation resistance of the whole material.

4. The combination should preferably have at least 5% ductility at room temperature, to make for ease of handling and fixing. This is undoubtedly the greatest difficulty to be overcome before the use of cemented carbides as blade materials becomes possible. The most suitable compromise between the ductility of the metal and the rigidity of the carbide is probably to be found in the use of equal proportions of each material, but the mechanical properties are then nearer those of the metal than the ceramic.

5. The unusual structure of the cemented carbides provides a very high damping capacity. This is a valuable characteristic, greatly reducing the susceptibility to fatigue failure.

Of the several special materials now in production the Turbide, WZ and Kentanium alloys are probably the most important. The composition and stress-to-rupture values for 100-hour tests are given in Table XI, together with data for certain other special metal-ceramic mixtures.

The Turbide alloys (manufactured by Hard Metal Tools Ltd.) are prepared from a solid solution of titanium and chromium carbides bonded with a nickel-chromium, cobalt-chromium, or nickel-cobalt-chromium matrix. During the sintering operation some of the chromium carbide is rejected from the mixed carbide phase and enters the metal to form a solid solution. The creep strength of such a material is found to increase with the amount of this dissolved carbide in the metal until saturation is reached, when a decrease in creep strength can be associated with the presence of free chromium carbide.

The WZ alloys are of Austrian manufacture and have similar properties to the Turbide materials. They are prepared by bonding titanium carbide with chromium and either nickel or cobalt. A certain proportion of the chromium enters the carbide phase during the sintering operation.

Kentanium alloys are made in America and have generally been prepared from a mixed titanium-tantalum-columbium carbide with a plain nickel or cobalt bond. The composition of newer and greatly improved alloys has not been released.

TABLE XI

ALLOY	COMPOSITION	SPECIFIC GRAVITY	STRESS TO RUPTURE IN 100 h (ton/in²)	TEMPERATURE (°C)
Turbide R.34	Medium Ni-TiC-Cr_2O_3	6·15	4·9	980
Turbide R.45	High Ni-TiC-Cr_2O_3	6·65	4·2	980
WZ.1b	60% TiC, 32% Ni, 8% Cr	6·20	6·3	980
WZ.1c	50% TiC, 40% Ni, 10% Cr	6·40	6·2	980
WZ.2	60% TiC, 28% Co, 12% Cr	6·10	4·6	980
WZ.12a	75% TiC, 15% Ni, 5% Co, 5% Cr	6·00	4·2	980
K.183A	80% TiC-TaC-CbC 20% Co	5·8	5·1	980
K.151A	80% TiC-TaC-CbC, 20% Ni	5·8	5·1	980
K.152B	70% TiC-TaC-CbC, 30% Ni	5·9	2·2	980
K.161B	—	—	6·9	980
K.162B	—	—	5·6	980
K.162C	—	—	5·8	980
K.163B	—	—	4·5	980
ZrC-Cb	87·5% ZrC, 12·5% Cb	6·29	—	—
B_4C-Fe	64% B_4C, 36% Fe	3·24	—	—
CrB-Ni	85% CrB, 15% Ni	5·9	1·16	827
Metamic LT-1	70% Cr, 30% Al_2O_3	6·0	—	—
Al_2O_3-Cr	70% Al_2O_3, 30% Cr	4·6	7·2	980
Nimonic 100	(for comparison)	8·04	4·2	980

At the present time experimental blades prepared from titanium carbide-nickel, titanium carbide-molybdenum, titanium carbide-iron and titanium carbide-Nimonic mixtures have been tested under engine conditions. The results appear to have been satisfactory provided steady temperature conditions are maintained. Fracture in the region close to the root fixing appears to be the commonest mode of failure, due to the stress concentration at this point. This difficulty may possibly be overcome by careful design.

In addition to the substances discussed above some attention has been given to pure oxides and mixtures of oxides and metals. Whereas conventional refractory materials, used in furnace linings, etc., generally possess a certain proportion of glassy material which flows under stress when hot, the pure oxides form intercrystalline

bonds which are almost as strong at high temperatures as at room temperature. Pure alumina, for example, has a tensile strength of some 18 ton/in^2 at 20°C and this does not fall to half the value until 1200°C (2192°F) is reached. This oxide, however, has a high coefficient of thermal expansion which renders it susceptible to damage during rapid heating or cooling and the complete lack of ductility makes it unusable where mechanical shock is possible.

The majority of the pure oxides are highly stable even at temperatures close to the melting point and are usually of low density compared with metals, as is shown by the following details:

TABLE XII

MATERIAL	APPROXIMATE MELTING POINT (°C)	SPECIFIC GRAVITY	THERMAL SHOCK RESISTANCE
Al_2O_3 ..	2050	4	Good
BeO ..	2500	3	Very good
MgO ..	2800	3·65	Good
ThO_2 ..	3000	9·5–10	Fair
TiO_2 ..	1850	4·1–4·3	—
ZrO_2 ..	2720	5·5–6·0	Fair

In the German wartime research programme a material called "Dug" was developed as a high-temperature alloy substitute and consisted of an equal mixture of alumina and iron powder, sintered together. There is no record of any actual use of this material however. A similar combination of alumina and chromium has been reported to have good oxidation resistance up to about 1500°C (2732°F).

Since ceramic materials are considerably stronger in compression than in tension, a German, Professor E. Schmidt, designed a turbine (2) in which the casing rotated around the stationary central portion. The blades were to be fastened on the inner surface of the casing and would consequently be under compressive centrifugal stress. The actual construction of such a turbine was not carried out before the war ended.

An experimental turbine at the Pamatrada Research Station is being used to investigate the possibility of producing wholly ceramic blades for operation with gas temperatures up to 1200°C (2192°F). The most promising material so far is zircon to which 2½% chromic oxide is added, which has been used for the manufacture of the stator blades. Trials have also been carried out with

blades made in a material manufactured by the Morgan Crucible Co. Ltd. and known as I.C. This is actually silicon carbide made by a special process. The blades are satisfactory in their resistance to the operating conditions and thermal shocks but tend to develop glassy areas on the surface in use.

Ceramic Coatings

During the war American research organisations produced several ceramic mixtures suitable for protective applications to metals exposed to high temperatures. The project was originally concerned with the exhaust supercharger fitted to aircraft such as the well-known "Flying Fortress". This unit was similar to a small single-stage turbine and received the hot exhaust gases from the internal-combustion engines. The heat of the gases was thus made to do useful work which would otherwise have been lost and the energy derived from the turbine provided a degree of supercharging. The ceramic coatings were required to provide protection for the metal forming the exhaust cones of these turbo-superchargers.

The production of such coatings calls for the matching of the properties of the ceramic mixture with that of the underlying metal, not an easy task. The requirements may be summarised as:

(a) High resistance to cracking and spalling under severe thermal and mechanical stresses.
(b) Protection of the underlying metal from oxidising and corrosive gases at high temperatures.
(c) Freedom from "blistering", which is often very troublesome with the conventional porcelains. This is associated with the occlusion of small air bubbles in the ceramic mixture, which expand during the subsequent fritting.
(d) Resistance to creep at the service temperature.
(e) Good adherence to the metal.

The results of this early work have been very valuable in the subsequent production of coatings for alloys, such as those based on molybdenum, which have high strength but no oxidation resistance. Alternatively, the ceramics may be used greatly to prolong the normal life of heat-resisting alloys, thus securing the maximum life from expensive materials, or to support the use of low-alloy materials, otherwise impossible because of their poor oxidation resistance.

In the gas turbine ceramic coatings will probably find their main application in the aircraft field, where frequent inspections are carried out and rapid replacement is possible should deterioration occur. The main disadvantage of such a treatment appears to be that, in order to provide a sufficiently protective coat, a considerable

thickness of ceramic is needed. A thick coating, however, adds to the weight of the component, is more difficult to hold in place and is more susceptible to thermal shock than a thin one. The successful development of ceramic coatings for the protection of flame tubes in aircraft engines has already been reported. This part of the gas turbine is subjected to high temperatures but is comparatively unstressed.

In industrial gas turbines, where the combustion chambers are generally on a much larger scale, there has been a tendency in recent years to incorporate a refractory brick lining in the combustion chamber. Special brick shapes have been developed to ensure that they interlock while permitting thermal expansion or contraction to occur. The use of a lining of this type ensures more perfect combustion, particularly in the case of the heavy fuel oils, and permits higher combustion chamber temperatures than would otherwise be possible.

Surface Treatments

In addition to the above method of surface protection there are several treatments which are applicable to alloys in order to induce special properties in the surface layers. These include cold-working, electrochemical treatments which produce a non-metallic coating (e.g. anodising) and heat-treatments which only alter the structure of the outermost regions. There are also several processes for altering the composition of the surface layers of an alloy, such as carburising and nitriding and the methods available for increasing the content of alloying elements, such as chromising, aluminising and siliconising. The latter are probably the most promising for use with gas turbine alloys, and brief details are given below.

Chromising

This metal can be introduced into the surface layers of low-carbon steels by treatment in either solid, liquid or gaseous media, each with particular advantages and disadvantages. The process is expensive but produces an extremely hard, corrosion-resistant surface.

The method used for many years consisted of heating the steel in a mixture of powdered alumina and chromium metal powder, in approximately equal proportions, to a temperature in the range 1300–1400°C (2375–2550°F). A hydrogen atmosphere was maintained in the container to ensure that the surface of the chromium particles remained free from oxide. The alumina played no part in the chromising operation but prevented fritting of the metal powder.

The depth of the "case" obtained depended on the time and

temperature of treatment but was usually between 0·0005 and 0·005 in. The chromium content at the outer surface varied from 10% to 30%.

The very high temperatures and long times required by this method, together with the rather rough surfaces produced, has led to the use of gaseous or liquid treatments. Both methods generally involve the production of chromium chloride and its subsequent decomposition at the surface of the steel. In the case of liquid treatments the components may be immersed in a bath of molten chromium chloride containing some diluent chloride, such as barium chloride, at 1100–1200°C (2000–2200°F). The most common method, however, involves the production of chromium chloride vapour, by the reaction of hydrogen chloride with chromium or ferro-chromium, and treating the steel for about 3–4 hours in the gas, suitably diluted with hydrogen.

The resulting chromium-rich surface layers are rather brittle and crack if plastic deformation of the steel occurs, but for applications where wear and corrosion resistance are required in the absence of relatively severe stresses, this process is very valuable. It is frequently used to improve the materials used for water turbine buckets, for example.

Siliconising

Increasing the silicon content of the surface layers constitutes a cheaper method of improving the resistance to corrosion and wear, particularly for use at slightly elevated temperatures. The case contains up to 14% silicon and although it is unmachinable it is more ductile than a chromium case. For the siliconising process low-carbon steels are preferable but high-carbon materials may be used, although the length of time required to produce a given depth of penetration will then be considerably greater.

The steel is subjected to the action of silicon carbide or ferro-silicon, in the presence of chlorine, at temperatures between 925°C (1700°F) and 1000°C (1850°F) for two or three hours. The depth of penetration may range from 0·005 to 0·10 in. An article treated by this method has a greatly prolonged life at temperatures up to 875°C (1600°F).

Calorising

The use of aluminium in various forms as a surface protection has been general for many years. The addition of this metal to the surface layers of a plain carbon or low-alloy steel produces a material highly resistant to oxidation or corrosion, particularly in the presence of sulphurous gases, for use at temperatures up to 760°C

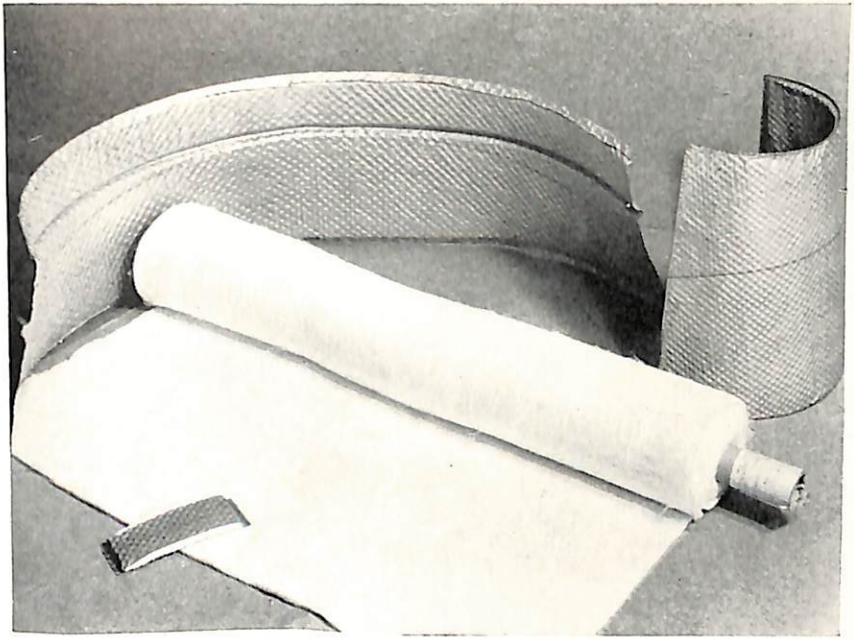


Plate 16. *Materials used in insulating blankets as manufactured by British Refrasil Co. Ltd*

Plate 17. *Insulating blanket in position on the afterburner of a General Electric J47 engine in an F-86D aircraft*

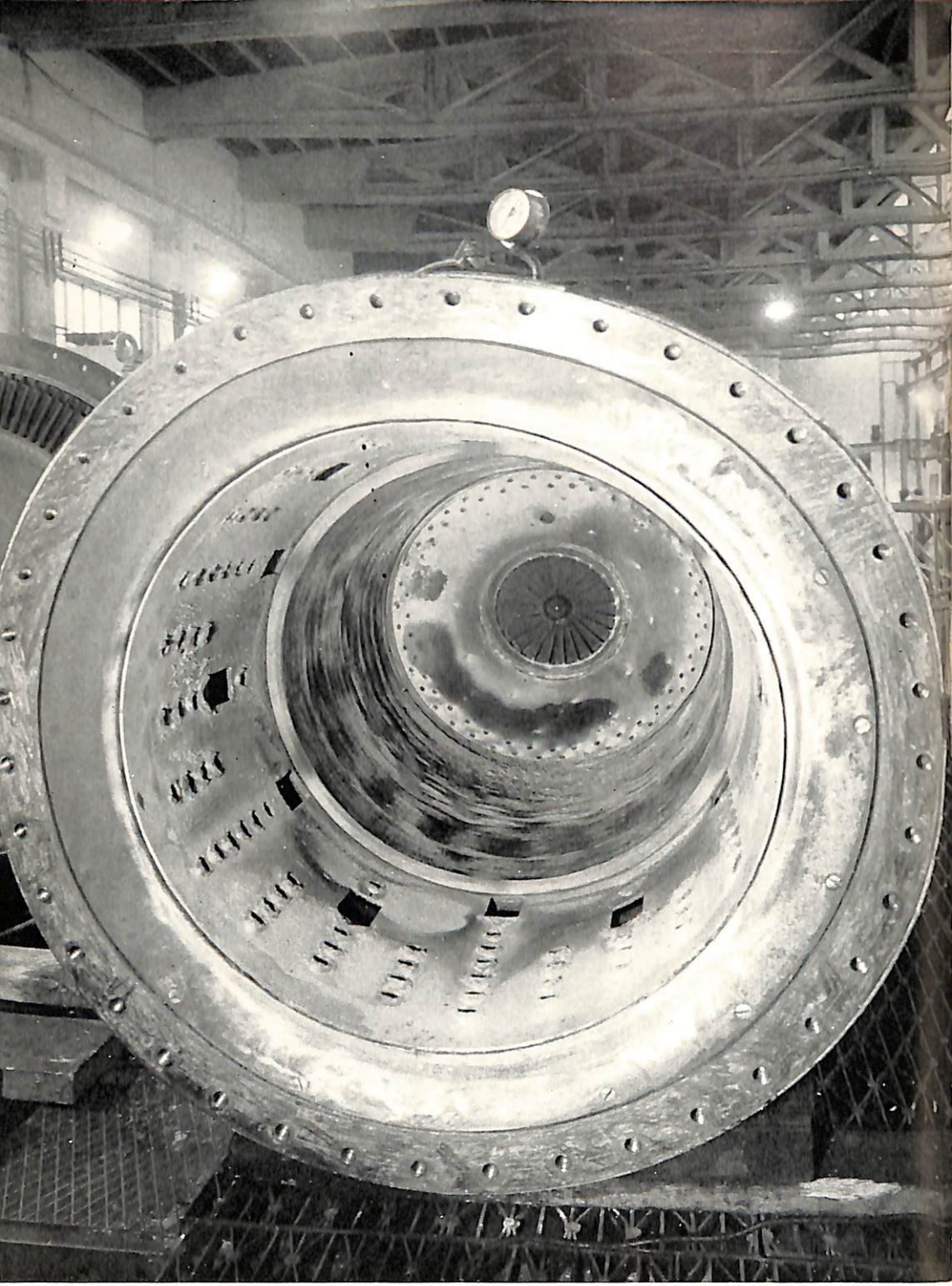


Plate 18. *An industrial gas turbine combustion chamber for a C. A. Parsons and Co. Ltd. unit*

(1400°F) or 925°C (1700°F), depending on the length of service required.

When steel is heated in a mixture of powdered aluminium and ammonium chloride in a neutral atmosphere, an aluminium surface content of 60% is possible. The temperature required ranges from 850°C (1550°F) to 925°C (1700°F) and the depth of penetration may be up to 0·006 in. The aluminium-iron alloy so formed is exceedingly brittle and the modern processes include a further treatment at 800–980°C (1500–1800°F) for periods up to 2 days, so that the aluminium diffuses further into the metal, reducing the maximum surface value to approximately 25% and increasing the depth of penetration to 0·025–0·040 in.

The final aluminised surface is reasonably ductile and permits about 5% elongation to occur before the development of cracks. Consequently the steel may be slightly hot-worked after treatment.

In some cases the use of special aluminium compounds brings about a combination of the impregnation and heat-treatment processes, thus appreciably shortening the total time required. Alternatively the aluminium may be introduced to the surface of the steel by hot dipping, followed by a suitable diffusion treatment.

Aluminised steel was extensively used by the Germans during the war to offset their serious shortage of strategic elements. It was used in various parts of their aircraft gas turbines, such as the combustion chamber and exhaust assemblies, the inferior properties being offset by suitable air cooling.

A process combining the deposition of chromium and aluminium in the surface layers has recently been developed and it is stated (3) that such a coating possesses oxidation resistance up to 1200°C (2200°F), as compared with 850°C (1550°F) for chromised steel.

Several other metals may be added by similar processes but the above constitute the main commercial treatments. One problem encountered in the introduction of these surface elements lies in securing uniform deposition, the areas at the bottom of depressions or grooves being particularly difficult. In addition to the above surface treatments there are several processes by which light alloys or ferrous materials may be treated to increase their corrosion resistance. Aluminium alloys for instance are greatly improved by anodising, in which a very thin film of alumina is formed on the surface, generally by electrolytic attack. Ferrous materials are particularly prone to rusting and corrosive attack and a much-used method for improving resistance to attack consists of producing a film of ferrous phosphate on the surface. Various names are associated with this process, such as Bonderising, Coslettising and Parkerising. The parts to be treated are immersed in a hot bath of a

dilute solution of ferrous phosphate and manganese di-hydrogen phosphate. This treatment is only applied to components not subjected to heat, since heat could destroy the film. It might be used, for example, on the outer surfaces of the compressor casings.

Clad Materials

The technique of rolling or pressing together two different alloys to give an otherwise unattainable combination of surface and core characteristics is a familiar one in the metallurgical field. This is known as "cladding" and usually refers to surface coatings where the thickness of the added layer constitutes more than about 5% of the total thickness. This surface metal may be added by any one of several methods according to the particular combination required, e.g. casting the cladding metal around a previously formed ingot, rolling composite layers to induce bonding by pressure welding, etc.

The most common engineering material produced by this type of process is probably stainless clad plate. This consists of mild-steel sheet covered on one or both sides by a thin layer of stainless steel and its use in place of stainless steel sheet greatly reduces the cost of equipment in which a high degree of corrosion resistance is called for.

In the gas turbine, clad materials have had little application until recently, when a material called Nimoply 75 was produced by Henry Wiggin Ltd., for combustion chamber linings, etc. This consists of two thin sheets of the oxidation resistant material Nimonic 75 with a sheet of copper between. The copper greatly improves the thermal conductivity and prevents the formation of "hot-spots".

Other methods of applying thin surface coatings include hard-facing, electroplating, spraying and dipping. Very little use is made of these treatments in gas turbine metallurgy at the present time.

Chromium and Molybdenum Base Alloys

The present precipitation-hardening heat-resisting alloys based on iron, nickel and cobalt require a solution treatment to be carried out at high temperatures and any working must also be carried out at very high temperatures. In addition the effects of thermal diffusion in weakening the alloys during service becomes more and more pronounced as gas turbine temperatures are increased. Consequently it is highly probable that 800–850°C (1500–1600°F) represents the maximum service temperatures for any precipitation-hardened alloys based on these metals under highly stressed conditions.

In the search for materials suitable for use at temperatures of 1000°C (1800°F), or even higher, the metallurgist is investigating the potentialities of metals with very high melting points. Of the many metals with melting points greater than 1600°C (2912°F), listed in Table XIII, chromium and molybdenum have been extensively studied. Of the other metals, the majority must be eliminated from economic or other considerations. When tests are carried out at high temperatures, a rough relationship is found to exist between the relative strengths of various materials and their melting points. It is thought that, since the melting point may to some extent be taken as a measure of the crystal bonding energy, it may be regarded as an indication of the hot-strength of polycrystalline metals.

TABLE XIII

Metals with melting points above 1600°C

	METAL	MELTING POINT (°C)	DENSITY
1.	Hafnium	1700	11·4
	Iridium	2454±3	22·4
	Masurium	2700 (calculated)	—
	Osmium	2700±200	22·5
	Platinum	1773·5±1	21·45
	Rhodium	1966±3	12·44
	Ruthenium	2500±100	12·2
	Rhenium	3170±60	20·0
	Protoactinium* ..	3000 (calculated)	—
	Thorium*	1800±150	11·5
2.	Columbium	2500±50	8·57
	Tantalum	3000±100	16·6
	Vanadium	1710±10	5·68
	Tungsten	3410±20	19·36
3.	Titanium	1820±100	4·5
	Zirconium	1750±70	6·4
4.	Chromium	1830±50	7·14
	Molybdenum	2625±50	10·2

* Radioactive.

Of these metals, those in Section 1 are too expensive to be considered or only available in very small quantities. Columbium,

tantalum, vanadium and tungsten are frequently added in small quantities to creep-resisting alloys but their use as major constituents appears very unlikely. All are highly susceptible to oxidation and the high density of tantalum and tungsten is a serious disadvantage for applications such as turbine blading where the majority of the stress arises from centrifugal forces. Titanium alloys have numerous low-temperature applications and are discussed in Chapter Four. Zirconium alloys are undergoing extensive development at the present time but are unlikely to find application in the gas turbine.

Chromium and molybdenum, however, have many properties which make them very suitable for use as the basic constituent of a high-temperature alloy. In each case there are also serious disadvantages which will have to be overcome before their application becomes feasible.

Chromium

A good deal of research on chromium and its alloys has been carried out in this country by the Fulmar Research Institute and in the U.S.A. at the Batelle Memorial Institute. The metal has a density of 7·14 g/cm^3 and the coefficient of thermal expansion is very low, $8{\cdot}1 \times 10^{-6}$ per °C. In addition it is highly resistant to oxidation and corrosion, due to the formation of a thin protective film of chromic oxide.

Research work (4, 5) has shown that chromium apparently possesses a serious inherent disadvantage which may prove insuperable. This is the fact that, although at room temperature pure chromium is soft, 90–100 Brinell, and readily sawn, it is extremely brittle. At red heat, however, the metal has a reasonable ductility.

This transition from the ductile to the brittle state takes place over a range of only a few degrees, the value varying with the purity of the sample, but lying above 50°C (120°F) for the purest material so far produced. Consequently at room temperature the metal and its alloys are extremely notch sensitive.

This property of chromium has been likened to the similar transition which is found in iron, tungsten and molybdenum. Fortunately the transition temperatures of iron and molybdenum are below room temperature. A study of the effects of impurities on the transition temperature for iron has shown that, although it is lowered by certain elements, e.g. manganese, an increase in the oxygen content materially increases the transition temperature. Similarly, in the case of molybdenum, it was not until the oxygen content of the metal was reduced to a very low figure (below about 0·003%) that a ductile form was obtained.

Investigations into the effects of impurities, especially oxygen,

in chromium have shown that apparently all elements increase the transition temperature to some extent and the amount of oxygen present has little significance. The lowest oxygen content so far achieved seems to be of the order of 0·001% and although this amount may possibly be shown to be sufficient to cause the transition temperature to lie above room temperature, it seems likely from the available evidence that the transition is an intrinsic property of chromium irrespective of the purity and may always lie above room temperature.

Since chromium crystallises with a body-centred cubic structure, there would appear to be no reason for this lack of ductility. Furthermore, there is good evidence (6) that, at very low strain-rates, chromium can behave in a ductile manner. Consequently the future experimental work on this metal must be concentrated on elucidating the reasons for this brittleness.

A further disadvantage in the use of chromium and chromium-base alloys devolves from the fact that, although the melting point is approximately 1830°C (3326°F), the boiling point is only some 350°C higher. Since vacuum melting is essential to prevent the rapid interaction of the metal with oxygen and nitrogen, this proximity of the boiling point means that a proportion of the charge is lost by volatilisation. In addition, chromium readily reacts with many refractory materials and melting must be carried out in crucibles of materials such as beryllia, zirconia or thoria.

These melting problems could no doubt be circumvented by the use of powder metallurgy techniques but, until the room temperature brittleness can be prevented, chromium can find little use as a high-temperature material.

Molybdenum

Molybdenum, with a melting point of approximately 2600°C (4700°F) has been used, in the form of ferro-molybdenum, as a strengthening addition in steels for more than a quarter of a century. The commercial production of the pure metal, however, is a more recent development, since its application is limited. High-molybdenum alloys, such as Hastelloy B, have however been developed and used in the gas turbine. These alloys are restricted to service at temperatures below dull red heat because of the excessive susceptibility to oxidation.

This rapid oxidation of molybdenum-base alloys is the only limiting factor in their application in the gas turbine. Manufacturing problems due to high resistance to deformation have been successfully overcome and the alloys have good creep-resisting properties up to 900°C (1650°F) together with sufficient room-temperature

ductility. In addition the coefficient of thermal expansion of molybdenum is only one-third the value of the majority of the high-temperature alloys.

The poor oxidation resistance is due to the high affinity of molybdenum for oxygen to form MoO_3, with a melting point of 759°C (1400°F). This oxide sublimes readily at temperatures of about 600°C (1100°F) and over, thus exposing clean metal again. This tendency cannot be suppressed by the usual method of adding chromium, since the amount of chromium required would result in an extremely brittle alloy.

Many attempts have been made to produce a suitable protective coating for these alloys. Three methods have been examined:

(a) Ceramic coatings.
(b) Metallic coatings, e.g. cladding.
(c) Surface treatment, e.g. siliconising.

Ceramic coatings have been successfully produced for protection at temperatures up to 1600°C (2900°F) but only for a relatively short time. The limit for lives of several thousand hours appears to be about 900°C (1650°F) at the present time and, protected in this way, molybdenum alloys have been used in experimental engines as combustion chamber linings, tail cones of aircraft jet engines, and nozzle guide vanes. The possibility that slight cracking or flaking of the coating may occur, which would lead to very rapid destruction of the component, is almost certain to prevent the use of these coatings in highly stressed conditions.

The cladding of molybdenum with heat-resisting alloys has also been reported and this method would seem to be preferable to the refractory type of coating. Little information is available on the properties of the composite materials so formed since they are still in the experimental stage. The mass production of blades from clad material would probably be a very complicated operation, however, since the coating would have to be perfectly gas tight.

Siliconising of molybdenum provides increased oxidation resistance to very high temperatures but the temperature for prolonged operation of such material would be restricted by diffusion effects and the tendency of the surface layers to flake under thermal or mechanical stress. The limiting factors in the use of the metals chromium and molybdenum, therefore, may be summarised as:

Chromium—complete lack of room temperature ductility in the metal and its alloys, making them exceedingly susceptible to thermal and mechanical shock stresses.

Molybdenum—susceptible to rapid oxidation when heated to temperatures above dull red heat.

Sintered Products

The well-established techniques which come under the heading of "powder metallurgy" are being used for the manufacture of experimental blades in which special properties are called for.

The basic principle of the powder metallurgy process consists of sintering together mixtures of metal powders. When particles of metal are brought into close contact under pressure, a certain amount of surface welding will occur. On heating to a sufficiently high temperature, diffusion across the interface occurs and the particles become firmly bonded together. In addition the lowered surface tension tends to minimise the surface area, so that if the heating is carried out under vacuum the metal particles will unite into a non-porous mass. When properly carried out the resulting product is indistinguishable from a melted material.

In this way, by mixing the required proportions of metal powders, any type of alloy may be made despite the fact that melting does not occur. Consequently the method is particularly applicable to the production of alloys with metals such as molybdenum, tungsten, etc., where the very high melting point makes melting and casting a formidable operation. Certain combinations of metals which do not alloy in the ordinary sense may also be prepared in this way, e.g. the silver-molybdenum mixtures used as electrical contact materials, although silver and molybdenum have such widely differing melting points, 960°C and 2625°C respectively, and are immiscible in the molten state.

Unless the inherent porosity of the material is completely removed, however, the mechanical properties of an alloy prepared in this way may be inferior to those of a cast or wrought product of the same composition. Consequently the application of this technique to standard alloys is restricted to the mass production of small components, where the high initial cost of the necessary dies may be justified.

An alternative to the development of materials to withstand higher temperatures in the gas turbine is now receiving attention and that involves the construction of hollow or porous components, which can then be cooled. Several methods are available for cooling the various parts of a gas turbine but the main type of cooling which would require the application of powder metallurgy techniques is that involving the passage of air or other fluid through pores in the alloy.

Although generally the elimination of porosity is one of the main problems in the manufacture of a component by powder metallurgy, the production of a porous material is not so simple as it would seem at first sight. This follows from the fact that, to be sufficiently

porous to allow free passage of air right through a certain thickness of material, the pores must be interconnected, while not being so numerous that the component is too weak for service requirements. In other words, the requirement of permeability is not necessarily fulfilled by a porous material.

The methods available for producing these special materials are still in the development stage. A very useful technique appears to consist of including a volatile material, e.g. ammonium compounds, in the metallic mixture. On heating, this substance gasifies and in escaping from the compact creates passages through it. The final permeability may be controlled by lightly rolling, when the diameter of the surface holes slightly decreases. To strengthen the bond between the metallic particles in a porous component it is sometimes preferable to pre-coat them with a relatively low-melting-point metal. On heating the effect of this coating is to "solder" the particles together, no actual sintering occurring in this case.

The limiting factor in the use of these porous materials, however, is likely to be the serious loss of mechanical strength which is caused by this type of structure. In an article (7) comparing the properties of a sintered alloy of the Vitallium type, i.e. 30% Cr, 6% W, 64% Co, with the very similar cast Haynes Stellite 23, it is shown that even when sintered to 98·6% of the theoretical density, the creep and fatigue properties at elevated temperatures are appreciably poorer, probably due to the absence of the grain boundary precipitation so important in creep-resisting alloys.

The oxidation resistance of a porous product is also likely to be a source of difficulty, since the surface area exposed is so greatly increased. This will be offset to some extent by the cooling effect of the air or liquid passing through the material.

Other Special Materials

In addition to the production of permeable components by the method mentioned above, a certain amount of development work is proceeding on the manufacture and testing of certain parts, such as turbine blading, complete with cooling passages. For many years hollow blades have been produced for use in the steam turbine by rolling a composite ingot of stainless iron with a mild-steel centre. After shaping the mild steel is dissolved out by acid, leaving a hollow blade shape of stainless iron. The production of a large number of small air holes through the wall in addition to the central hollow region presents a more complex manufacturing problem, but there are almost certain to be major developments in this field in the near future.

A turbine built for the National Gas Turbine Establishment by

the General Electric Co. Ltd., uses air-cooled blades produced by powder methods. One such blade is pressed in a powdered Vitallium-type alloy and incorporates some 40 or 50 thin cadmium wires along its length. A low-temperature heat-treatment removes the cadmium by volatilisation before sintering is carried out, leaving channels for air to pass through during service. This turbine also uses extruded Nimonic blades, with cooling passages incorporated. These blades are discussed later.

REFERENCES

1. J. H. Westbrook. *Bull. Amer. Ceram. Soc.*, June, July, 1952, Vol. 31, 6, 7, pp. 205–8, 248–50.
2. B.I.O.S. *Overall Report No.* 12, 1949, H.M.S.O.
3. D. M. Dovey, I. Jenkins and K. C. Randle. *Symposium on Properties of Metallic Surfaces*, Institute of Metals, 1952.
4. A. H. Sully and others. *J. Inst. Met.*, 1952–53, Vol. 81, pp. 569–8.
5. R. M. Parke and F. P. Bens. *Symposium on Materials for Gas Turbines*, A.S.T.M., Philadelphia, 1946, p. 80.
6. H. B. Goodwin, R. A. Gilbert, C. M. Schwartz and C. T. Greenidge. *J. Electrochem. Soc.*, 1953, Vol. 100, p. 152.
7. R. W. A. Buswell, W. R. Pitkin and I. Jenkins. *Iron and Steel Institute Special Rep. No.* 63, p. 258.

SIX

The Manufacture of Gas Turbine Components

THE RIGOROUS service conditions of many parts of the gas turbine demand a high standard of production in the manufacture of its components. In this type of engine it is not surprising that almost every method of working and shaping metals should be used at some stage of its construction.

There are many parts, however, which are neither highly stressed nor subjected to high temperatures and the manufacture of such parts presents little or no difficulty. The components calling for the highest standards of quality and reliability have already been mentioned in Chapter Four in discussing the selection of an alloy for such components. In this chapter the various production techniques used in their manufacture will be considered.

THE COMPRESSOR

The method of manufacture of the compressor disc and blading will vary according to whether an axial or centrifugal design is chosen and also with the alloy to be used. The centrifugal compressor can be most readily produced either by casting, forging or by a combination of casting and machining. With the majority of the light-alloy compressors used in aircraft gas turbines it is possible to produce disc and blading as an integral casting or drop-stamping, requiring only a final machining to size. Alternatively it is sometimes preferable, in the case of more complex designs, to mill the impeller blades from the rough casting as straight pieces, subsequently bending them to the required shape by special machines and welding them to the disc in the correct position.

In the case of the axial compressor, where the stronger alloys are generally used, the blades are usually produced individually by one of the following methods:

(a) Investment casting to close tolerances, possibly followed by surface finishing operations.
(b) Machining from bar stock.
(c) Production as either wide limit or precision forgings from bar, followed by final machining operations if necessary.
(d) In certain cases the blades may be machined as an integral part of the disc.

In the first three methods the blades are subsequently assembled

on the disc by one of several techniques. Method (c) is probably the cheapest but all, with the exception of (d), which is only feasible with simple aerofoil sections, are in general use.

Combustion Chamber and Auxiliary Ducting

The metal combustion chamber, as opposed to the refractory brick type, is a combination of an outer shell, generally of sheet steel, and an inner flame tube which is made from a heat-resisting iron- or nickel-base alloy in sheet form. The thickness of the metal is kept to the minimum consistent with the strength requirements and is usually about 0·015 in. Consequently it is possible to use the standard sheet-metal techniques, such as the combination of bending and longitudinal welding, or pressing and deep-drawing for weld-free constructions. The latter has several advantages from both the production and service aspects, but in view of the limited ductility of many heat-resistant alloys may involve a large number of operations. The use of welded joints, on the other hand, frequently causes trouble under operating conditions, due to the poor behaviour of the weld under stress as compared with the body of the material. It is certain that, as combustion chamber design improves and the pressing of these special stiff alloys becomes more common, the amount of welded construction will decrease.

Turbine Discs and Rotors

The varied mechanical strength requirements of the disc and rotor materials have already been discussed. From the manufacturing aspects the main considerations are the production of sound, clean ingots followed by careful forging and heat-treatment and possibly the application of special techniques, e.g. that known as "warm-working".

The difficulties experienced in the shaping of the rotor assembly naturally increase with the size required, especially where the discs are to be forged integral with the shaft and it may be found preferable in the future to produce more rotors of the built-up type as the technique improves. Welding joints of a very high standard are called for in such cases.

Where separate discs are chosen, these are forged roughly to shape, including a short stub shaft, followed by machining to the required dimensions. The flash butt method of welding such discs to the rotor shaft has been recommended in preference to bolted joints, to obviate stress-concentrations at the bolt-holes.

Turbine Blading

Similar production methods to those already outlined for the axial compressor blading are used for turbine blading of both the stator

and rotor types. In general, machining from rolled bars of special section or rough stampings is now the accepted method in this country for rotor blading, while the new precision-casting technique has so far been restricted to the lower-stressed stator blades. After machining, finishing operations such as grinding and polishing are used to provide the high surface finish required to minimise fatigue failure. The developments at present taking place in producing blades with passages for cooling air have led to the use of a special extrusion technique. It seems likely that extruded blades will be used in the very near future so that some consideration of extrusion is required.

Heat-exchangers

With the exception of the casing, the whole of the usual type of heat-exchanger is built up from tubing with welded joints where necessary. Since the majority of the alloys used are appreciably stiffer than the usual tubing alloys, the manufacture of heat-exchanger tubing requires higher working temperatures, more powerful machinery and frequent intermediate annealing treatments to offset the work-hardening. The wall thickness of the tube influences the rate of heat transfer from the waste gases to the cold air and has to be as thin as possible while retaining a sufficient degree of high-temperature strength.

Basic Methods of Component Production

The basic methods of production involved in the manufacture of gas turbine components may be summarised, therefore, as:

1. Melting and ingot production.
2. Forging and rolling, possibly to special shapes.
3. Casting, by either general or precision methods.
4. Sheet metal techniques.
5. Machining, grinding and special finishing operations.
6. Extrusion.
7. Welding.
8. Heat-treatment.
9. Annealing, pickling and de-scaling.
10. Inspection and testing.

Melting and Ingot Production

High-alloy steels and special alloys are generally melted in electric furnaces, since these possess several features which make them specially suitable for these materials. The melting points of many of the alloying additions are so high that higher furnace tempera-

tures than are practicable with gas or oil-fired furnaces are required, whereas electric furnaces, though requiring a more costly source of power, allow rapid melting to be carried out. The effect of chromium in reducing the fluidity of a melt is particularly marked in heat-resisting alloys, where the chromium content may sometimes be as high as 30%, and consequently the pouring temperature for such a melt is much higher than that required for chromium-free alloys. The much smaller volume of slag required in electric steel melting also ensures that the final ingot is relatively free from inclusions which could, in service, act as serious stress-raisers.

Certain types of electric furnace also allow melting to be carried out under a vacuum or a protective atmosphere, such as argon, so that the addition of alloying elements which are liable to oxidise rapidly at the high temperatures involved can be carried out without serious wastage.

The special problems associated with furnace heating by electricity, however, limit the amount of alloy which can be made in a single melt and it is usual for such furnaces to be limited to outputs from 5 to 25 tons in the majority of cases.

The types of furnace favoured, both for the actual alloy production and any subsequent re-melting by casting manufacturers, vary with the weight of charge required, the type of alloy and the use to which the molten metal is to be put. Induction heating is necessary for vacuum melting or melting carried out under protective gases. Furnaces using this method of heating are particularly valuable where a very low carbon content is called for and also where troubles from inadequate mixing of the molten metals may be anticipated, since considerable agitation of the melt is set up by the induced currents.

Carbon arc furnaces, generally of the indirect arc type, provide a reducing atmosphere due to the consumption of oxygen by the arc. There is, however, the disadvantage that, either from pieces of electrode dropping in the bath or by reaction with the carbonaceous gases, there is a tendency for the carbon content of the charge to rise appreciably during melting. The minimum carbon content attainable in these furnaces is probably of the order of 0·05%.

The addition of alloying elements in the pure state to the furnace is generally undesirable since, if the melting points are above the furnace temperature, the process of melting and solution would probably extend over a considerable period and to avoid this it is usual to add them in the form of a ferro-alloy. This is clearly only possible where the iron content of the alloy is also high and if this is not so some other diluting metal must be chosen. Wherever feasible, however, iron is preferable since the ferro-alloys are freely

available. Chromium, for instance, is available as low-carbon, high-carbon, or high nitrogen ferro-chromium, as shown below:

Low-carbon ferro-chrome ..	0·02–2·0% C, 28–33% Fe
High-carbon ,, ,, ..	4·5–7·0% C, 30–35% Fe
High-nitrogen ,, ,, ..	0·2–5·0% C, 0·75% N, 30–40% Fe
Ferro-silicon	Available in a wide range of compositions
Ferro-vanadium	35–50% V, with up to 10% Si
Ferro-tungsten	80% W
Ferro-zirconium	10–40% Zr, with 40–50% Si
Ferro-columbium	50–60% Cb
Ferro-molybdenum	55–75% Mo
Ferro-titanium	20–45% Ti, 0·1% C; 15–20% Ti, 3–8% C

The temperature of the molten metal has to be carefully controlled for the casting operation since, if too high, a large grain size accompanied by excessive segregation may result while, if too low, an unsound ingot will be produced due to inadequate time being available for proper shrinkage and feeding to occur.

The highly alloyed steels and certain non-ferrous alloys, by virtue of their composition, are liable to undergo air-hardening if the ingot is cooled too rapidly after solidification. In such cases it is advisable to anneal the ingots while still hot or cool slowly in pits to avoid cracking. Even where air-hardening is unlikely, however, slow cooling is preferred since the inherent high resistance to deformation of these alloys might lead to the formation of internal cracks if excessive thermal stresses are set up.

The main faults likely to develop in ingots are excessive piping, segregation, blow-holes and inclusions. The formation of a pipe, or contraction cavity, is unavoidable since the volume of the solid alloy is always less than that of the molten materials. The ingot mould is so designed, however, that the cooling of the top part of the metal is slightly slower than the lower portion. In this way contracting metal can draw on the reserve of molten material. Even in well-designed moulds some pipe is formed but it is restricted to the upper portion of the ingot and the piece containing it is cut off at an early stage of the cogging or forging process.

Segregation results from the different melting points and solubility values of the alloying elements. As the cooling proceeds, certain substances, such as carbides and nitrides, come out of solution and, depending on their density relative to that of the molten

metal, may either rise or fall through the liquid portion of the ingot. The more soluble elements tend to concentrate in the centre of the ingot since, when alloys in a system having an extended freezing range are cooling, there is always a tendency for the higher-melting-point component to crystallise out first. The remaining liquid thus contains a greater and still greater proportion of the second component until a temperature is finally reached where solidification is complete. This leads to a type of cast structure known as "dendritic". This tends to be broken-up during the subsequent working of the ingot though the process is rarely complete and it is quite usual to find clear traces of the original structure after very extensive reduction.

Dissolved gases in the molten metal are evolved during cooling since their solubility decreases rapidly as the freezing point is approached. These gases form small cavities in the ingot, if there was insufficient time for them to be completely expelled from the metal before complete solidification. Fortunately the majority of these blow-holes are usually sealed up during rolling since the walls of the cavity are generally clean and weld together easily.

Hydrogen is a very troublesome gas to the steelmaker. From a variety of sources, particularly moisture in the furnace charge and the atmosphere, hydrogen is formed during the melting operation and becomes dissolved in the liquid steel. The solubility in the solid metal is quite high at temperatures near the melting point so that there is little or no tendency for the hydrogen to leave the steel during solidification. As the solid steel cools, however, the hydrogen comes out of solution and diffuses out of the metal. If the amount of hydrogen remaining when the temperature falls to about 100–200°C (210–390°F) is above a certain level, a severe form of internal cracking will result. From its appearance this is known as hair-line cracking.

The minimum hydrogen content for cracking seems to vary from steel to steel, nickel steels being apparently most susceptible, and it may also be related to the cooling stresses in the steel. While there is still much to be learned about hydrogen in steel, it is thought that the combination of the decreased solubility and lower rate of diffusion results in the hydrogen collecting at lattice defects in the metal under such very high pressures that cracking occurs.

Where subsequent cogging and rolling will be carried out there is generally no objection to the formation of hair-line cracks in the ingot, since they will weld up again during the deformation. In the case of large forgings, however, where only a relatively limited amount of working will be carried out, it is necessary to keep ingot and forging continuously at a fairly high temperature, sometimes

for periods of several weeks, until the hydrogen content has fallen to a sufficiently low level.

A recent development in Germany by the Bochumer-Verein Ag. may in some cases reduce the prolonged heat-treatment given purely to avoid hair-line cracks. It consists of vacuum treating the molten steel so that the hydrogen content is greatly reduced at the earliest stage. Three techniques have so far been used, depending on the quantity of metal involved. For casts of a few tons only either the ladleful of steel or the freshly poured ingot is subjected to the vacuum in a special tank equipped with very rapid-acting pumps. For casts of 10 tons or over, however, the surface area available in such conditions is small in relation to the weight of metal and the time required to remove most of the hydrogen is longer than that taken by the steel to solidify.

For these quantities, therefore, a special ladle arrangement has been developed. The molten steel is tapped from one ladle to another under vacuum, thus ensuring that the very large surface area of the stream of metal is exposed for treatment. It is important, of course, that sufficient super-heat should be given to the steel to ensure that after this treatment it will still be hot enough for successful casting.

Inclusions are non-metallic substances generally formed by the constituents of the alloy during melting and entrapped during the casting operations, e.g. sulphides, silicates and oxides. Particles of such materials do tend to rise to the top of the molten metal on standing but the time required is generally longer than the cooling of the alloy allows, so that care must be taken to prevent excessive mixing of slag and metal in the ladles.

Forging and Rolling

The production of bars or special sections in heat-resisting alloys is very similar to that involved in the case of ordinary steels. The ingots are generally machined to prevent surface defects in the finished bar, and, in the case of small billet sizes, surface dressing will also be carried out at an intermediate stage of rolling. Very stiff alloys may have to be re-heated at intervals during rolling to offset the greater increase in the resistance to deformation as the metal cools. If rolling is carried out at too low a temperature severe cracking of the bar may occur while in certain steels the use of too high a temperature will produce a greatly enlarged grain size. The re-heating is carried out in special furnaces where neutral or reducing atmospheres are maintained to prevent excessive surface deterioration of the alloy. Accurate temperature control is pro-

vided to ensure that the metal is heated to the highest possible temperature without risk of overheating.

After rolling to bar, the alloy is mechanically or chemically cleaned to remove the scale layer and expose the surface for examination.

Forging

The forging operations used in the rough forming of parts fall into four general categories:

(a) *Flat or Open Die Forging*, using the anvil and steam hammer technique.

(b) *Drop Forging*, in which the metal is shaped between two steel dies, one fixed to the hammer and one to the anvil. The die is designed to conform in size and shape to the finished article with an allowance for the cooling contraction and subsequent final machining. In addition, it is necessary to taper all the die inserts to make provision for the squeezing out of excess metal. This method is the one most widely used for the manufacture of small forgings. A steam hammer is generally preferred because of the greater control that can be exercised and the more rapid speed of operation.

(c) *Upset Forging*, a rapid production method generally used for small components such as bolts, in which a bar is held between grooved dies and pressure brought to bear axially. Under the applied force the end of the bar is enlarged to a shape controlled by the die design, or "upset" as the process is termed.

(d) *Press Forging*, a variation of the drop-forging technique in which all the shaping is done in one movement instead of several. This method is preferred in the case of large components since the use of steady pressure has a greater influence in working the metal right through to the centre of the cross-section than can be obtained by a number of rapid blows.

The application of these forging methods is usually restricted to relatively simple shapes and in addition the selection of this method of shaping will clearly depend on the forgeability of the alloy. Many of the creep-resistant alloys are so resistant to deformation that drop or upset forging to close limits becomes very expensive, due to the short die life obtained.

The term "precision forging" is sometimes applied to the method by which bar stock is first machined, upset or rough forged to the approximate shape followed by pressing in suitable dies to the final dimensions. In this method the deformation required in

the die is greatly reduced and the finished product only requires a surface cleaning to remove the scale formed. It is particularly applicable to the production of compressor blades in materials such as aluminium bronze, Monel, aluminium alloys, etc.

The temperature of the material for forging will depend on the composition but, with the exception of aluminium and copper-base alloys, the majority of the creep-resistant alloys and alloy steels are normally forged in the range 900–1300°C (1650–2400°F). In addition certain special operations below this range of temperature are sometimes applied to rotor and disc alloys in order to increase the proof stress. This technique, generally referred to as "warm-working", consists of a small controlled deformation in the temperature range 600–900°C (1100–1650°F), using any of the usual forming operations. The amount of deformation and the temperature required varies according to the composition of the alloy and the properties required in the component.

Where materials are to be subsequently heat-treated or used at elevated temperatures, excessive cold-work is to be avoided since it may tend to produce an unusually large grain size during recrystallisation. This will be particularly true where non-uniform sections are concerned since there is then a greater possibility of some part receiving the critical deformation necessary for abnormal grain growth.

Castings

Many components of the gas turbine are produced by re-melting alloy stock in small electric furnaces and casting the molten metal directly into moulds having the shape of the required part. The majority of rough castings are made in sand or, in the case of the low-melting-point light alloys, metal moulds. Where higher quality castings are required to closer dimensional limits, the process may be adapted to include oscillating moulds or the application of high-pressure air, both these requirements helping to ensure complete filling of the mould. The use of these types of casting in the production of gas turbine components is limited to those not subjected to high temperatures or stresses, such as the compressor casings, etc.

A special casting technique which has been widely used for the production of gas turbine components is that known as centri-spun die casting. This process involves the very rapid rotation of the mould during the pouring and cooling of the molten alloy. The rotary motion throws the metal outwards and the subsequent solid product has superior properties to the stationary casting.

Two types of machine are used, one rotating on a horizontal axis and the other vertically. The choice of method is mainly governed

by the size and design of the part to be produced, though there are no hard and fast rules in this respect. Generally, components with an overall length greater than the radial thickness are cast by the horizontal method while, if the reverse is true, the vertical machine may be preferred.

Among the several advantages claimed for the centri-spinning process, the main ones may be summarised as:

1. An ability to conform to complexities in design, since the centrifugal forces assist the liquid metal to penetrate to all parts of the mould.
2. Greater freedom from shrinkage cavities and porosity. As a result of density differences, slag particles tend to be centrifuged to the inner region of the casting, where their influence on the properties of the casting will generally be least, or may be more easily machined out if necessary.
3. Absence of directional properties.

The material used for the construction of the mould may be either refractory or a special alloy. The use of a refractory mould, however, is liable to introduce a higher proportion of inclusions into the metal. When the metal is first put into the mould it does not move with the same speed and consequently there is a considerable eroding action on the wall until the metal achieves the same speed of rotation as the mould. This erosion will tend to detach refractory particles from the mould wall and the time required for complete solidification will be insufficient for these particles to pass through to the inner surface of the casting.

In addition to the erosive action, a refractory mould involves a slower rate of cooling than the metal-walled type which may result in a coarser-grained casting.

Among the many castings now produced regularly by the centri-spinning process are such parts as nozzle shroud rings and diffusers.

In the production of blading the recently developed "precision casting" process is frequently used. It is largely restricted to compressor blading and turbine nozzle guide vanes in this country at the present time because of the general preference for forged rotor blades, but has been used for these latter components in the U.S.A.

The method is a modern version of a very old-established practice, the "cire perdu" or "lost-wax" process. This method was known to the Chinese and Egyptians over 2000 years ago and in the intervening period it has often been used in the reproduction of statuary. In more recent years the dental profession have used it extensively. During the search for more rapid production methods during the last war the process was intensively developed and has now become

a standard technique for the large-scale manufacture of castings of every description. The term "precision" generally applied to the method may be somewhat misleading but with careful control the majority of the castings can be turned out with a tolerance of two or three thousandths of an inch, which is "precise" when compared with the accuracy attainable by other casting methods. A more satisfactory description is obtained by the use of the term "investment casting".

The method is extremely useful in the production of complex designs not easily accommodated by other methods and for the forming of parts in alloys which are impossible or very difficult to forge. At the present time the use of the method is restricted to the production of articles with a maximum weight of about five pounds, this limit being imposed mainly by the loss of accuracy on larger castings and the greatly increased refractory troubles associated with supporting such a large mass of metal without the possibility of distortion or fracture of the investment. Except in the case of alloys which are very difficult to machine, the process loses its economic advantages as the size of the casting increases and it is probably in the two or three pounds' weight range that the advantages of investment casting are at a maximum.

Briefly the process consists of the following stages:

1. Preparation of a die from the master pattern.
2. Production of wax models from the die.
3. Coating of the wax model with a refractory "slurry" which dries to form a hard skin.
4. Reinforcing of the fragile ceramic coating with a surrounding layer of other refractory material.
5. Melting and burning out of the original wax model, leaving an accurate refractory mould.

It will be appreciated that the final accuracy of the casting will depend to a considerable extent on the original master pattern. This does not simply reproduce the dimensions of the required component but must make allowance for many factors, such as shrinkage of the wax pattern and the metal.

From this master pattern the dies are produced in either rubber or soft metal. The rubber dies are cheaper but are only used where the highest dimensional accuracy is not required. The soft metal dies are made from a variety of low melting-point alloys based on tin, lead, zinc, bismuth or similar metals and are more accurate and durable. Where a very long life and high precision is called for, low alloy or tool steel dies may be used but in such cases the master

pattern must be dispensed with, the required shape being machined directly from a drawing.

The molten wax is usually injected into the die in a semi-fluid condition in order to avoid thermal swirling effects and in many cases air pressure is applied to ensure perfect filling.

After solidification and removal from the mould the model may be either used immediately or stored for some time. To prevent distortion the rooms are maintained at a uniform temperature and humidity.

The composition of the wax used varies from manufacturer to manufacturer, but the melting-point is usually in the range 60–80°C (140–175°F). The shrinkage on solidification should be as low as possible and this can be brought about by blending the various types of wax. Since the wax has to be melted out at a later stage it is important that it shall be free from impurities, and that any residue remaining on the mould walls will burn away without leaving any carbonaceous or other residue.

The wax patterns are generally sprayed with the refractory coating which will constitute the face of the final mould, but dipping may be used. The properties of this coating determine the surface finish obtainable on the casting and careful attention has to be paid to the particle size distribution. During the casting process this coating has to be sufficiently strong to resist the heat and erosive effects of the molten metal but at the same time a certain degree of porosity is desirable to allow the escape of gases liberated from the solidifying metal.

In order to combine the superior surface finish obtained with the use of finely powdered material with the porosity of coarse particles, two or more coats may be applied. The choice of refractory is dependent on the alloy to be cast but is usually sillimanite, silica, alumina, zirconia, or mixtures of these materials. It will be appreciated that interaction between the hot metal and the refractory would give a very poor surface finish and is to be avoided.

After careful drying of the primary coating the wax patterns are arranged in suitable groups in a metal box, with wax patterns added to form the feeder and riser channels. The box is then filled with a "slurry" of a secondary investment material. This usually consists of crushed firebrick, quartz, zircon, or sillimanite, etc., again carefully graded. The slurry contains a bonding agent, such as ethyl or sodium silicate, and on vibrating the mould box for some hours the refractory is effectively packed around the patterns, the surplus liquid being drained away. The box is then heated for some hours at 40–50°C (105–120°F) to complete the drying operation and the temperature subsequently increased to 120°C

(248°F) when, on inverting the box, the wax drains out. Finally the refractory block is heated very slowly to about 1000°C (1830°F) to complete the hardening of the material and to remove the final traces of wax. The metal is then cast into the mould at that temperature since this helps to ensure accurate reproduction of surface details and proper running of the metal, especially in the designs involving thin sections.

After solidification the castings are carefully knocked out of the refractory block and sand or shot-blasted. Visual and more special methods of examination, which will be discussed later, eliminate any defective castings. The metallurgically sound castings are then checked for dimensional accuracy. The permissible variation will clearly be governed by the use to which the component is to be put and the percentage of acceptable castings will depend on this tolerance but the latter generally ranges from about 60% at ±0·002 in. to about 95% at ±0·005 in., for high-temperature alloys.

Ethyl Silicate

This compound came into prominence with the development of the precision casting technique, and is probably the most frequently used binding agent in this process. It is produced by the interaction of ethyl alcohol and silicon tetrachloride, the latter being formed by passing chlorine gas over heated silicon:

$$Si + 2Cl_2 \rightarrow SiCl_4$$

$$SiCl_4 + 4C_2H_5OH \rightarrow \underset{\text{ethyl silicate}}{Si(OC_2H_5)_4} + 4HCl$$

When ethyl silicate is mixed with water, hydrolysis occurs:

$$Si(OC_2H_5)_4 + 4H_2O \rightarrow Si(OH)_4 + 4C_2H_5OH$$

This reaction takes place rapidly in alkaline solutions but only very slowly in the presence of mineral acids. A typical stock solution for addition to the refractory materials in the slurry would comprise:

500 ml ethyl silicate
100 ml dil. hydrochloric acid (0·5%)
1400 ml alcohol

This is added to the slurry together with a small amount of magnesium oxide. The basic oxide neutralises the mineral acid and the hydrolysis is accelerated so that it is completed during the period of mechanical vibration. The alcohol is driven off during the heating at 40–50°C (105–120°F) and the final heating to 1000°C (1830°F) converts the hydroxide to the oxide:

$$Si(OH)_4 \rightarrow SiO_2 + 2H_2O$$

The silica so formed acts as a firm bonding agent and the micro-cracks produced by the evaporation of the alcohol provide the required degree of permeability.

In addition to the use of wax, other materials have been adopted, including plastics and mercury.

The use of plastic pattern materials, e.g. polystyrene, requires much higher injection pressures and temperatures than those used for wax. Although widely used they have certain limitations which may be summarised as:

(a) Greater tendency to distortion of the green investment during the melting-out operation.
(b) Flashes formed on the patterns during moulding are more frequent due to the greater pressures used, and have to be carefully removed.
(c) Higher cost of the plastic, generally coupled with the inability to re-use after melting out.

Plastics do, however, tend to allow a higher rate of production to be maintained and may be more readily stored for long periods without troubles due to distortion or deterioration.

The Mercast Process, developed by E. F. Kohl, is now widely used for the mass production of precision castings in the U.S.A. Advantages of using mercury instead of wax include greater dimensional accuracy, a superior surface finish and a very high quality in reproduction of detail. A further advantage of this material lies in the fact that two pieces of frozen mercury may be readily welded together by slight pressure and consequently the construction of intricate patterns is greatly simplified. Probably the greatest disadvantage of this process is the very high initial cost of the mercury required, since several thousand tons may be in circulation, but the wastage is very small.

The method is similar in outline to the lost-wax process. A machined steel die is essential for producing the mercury patterns, since this material does not deteriorate in use as rapidly as soft metal. This, of course, is a costly operation but the extra expense is offset by the much longer die life obtained.

The die parts are assembled in a tank containing acetone and the mercury poured in. The acetone forms a very thin film between the mercury and the steel, thus ensuring that sticking will not occur after freezing. The assembly is now transferred to the freezing-tanks, containing acetone cooled to a temperature of $-60°C$ to $-75°C$ ($-75°F$ to $-105°F$). The mercury rapidly freezes, the melting point of mercury being $-38{\cdot}85°C$ ($-38°F$), and is removed from the die and transferred to further refrigerated tanks containing

a ceramic suspension. Several successive dips are required to build up a satisfactory thickness of the refractory on the surface of the pattern before a careful evaporation treatment is applied to remove excess liquid. The invested pattern is then allowed to warm back to room temperature under carefully controlled conditions when the mercury melts and flows cleanly out of the refractory shell.

Since mercury undergoes a 3% volume expansion on passing from the solid to the liquid state, care has to be taken to prevent any of the metal in the thinnest sections of the pattern melting before the outgoing channel is free. Otherwise distortion or cracking of the refractory would occur. This is prevented by pouring warm mercury over the feeder section of the mould, thus ensuring that this part is melted first. In the process no troubles arise from the incomplete removal of the pattern material since the tendency of the mercury to stick to the walls of the investment is so much less than that of wax.

The moulds in this process are frequently used without the further addition of a secondary investment material, the $\frac{1}{8}$-in. thick wall built-up in the 10 or 20 successive dips being sufficiently strong after firing to hold small weights of metal. Otherwise, however, the process continues on the same lines as those given for the lost-wax process.

A further process, generally known as "shell moulding", is also being employed to an increasing extent for the production of many of the castings for use in the gas turbine. It has also been used for the production of nozzle guide vanes.

Details of this new method of sand moulding were first published in 1947. It originated in Germany during the war as the "C" or Croning process and the technique was subsequently described in the reports of American technical investigators.

Shell moulding is a technique for the manufacture of castings in a mould which is only a thin shell of refractory, approximately $\frac{1}{4}$-in. thick. The rigid bonding of the refractory particles is due to the incorporation of a small amount of thermo-setting plastic. As in the lost-wax process the main mould is subsequently supported by low-grade refractory material during casting.

Briefly the method consists of placing a heated metal pattern in contact with a quantity of the refractory and plastic mixture. According to the time of contact and the heat in the pattern a suitable thickness of refractory adheres to the metal pattern. After a suitable curing treatment the finished mould can be removed from the metal pattern, assembled in a suitable box with a supporting material and the molten metal poured in. Despite the fact that the plastic is burned off by the hot metal the mould retains its

shape long enough for the casting to solidify. There is, however, a tendency with steels for a certain amount of carburisation to occur at the surface and this may be troublesome with certain types of stainless steels.

The shell moulding process is most useful in the mass-production of high-quality castings where the expense of precision casting is not justified. The dimensional accuracy is, however, quite good and machining allowances may be greatly reduced or even eliminated. Since the amount of refractory used is much less than in ordinary sand casting a much closer control of particle size, etc., is possible and an excellent surface finish is thus obtained. Generally the castings need little or no surface cleaning.

Sheet-metal Forming

The extensive use of sheet metal for the construction of several parts of the gas turbine is made possible by the development of alloys having sufficient high-temperature strength for the service requirements while retaining a considerable degree of ductility. Typical components produced from sheet include the flame tubes, combustion chamber, ducting, etc. The large quantity of thin-walled tubing required in the heat-exchanger may also be considered under this heading, since the mechanical properties of tubing alloys are in many respects similar to those of sheet metals.

Sheet metal for use in extensive forming operations has to be of a high quality and free from surface and internal defects. The billet is hot-rolled to slightly oversize, pickled to remove the scale and finally cold rolled through highly polished rolls. This treatment provides a greatly improved surface finish and higher accuracy in rolling to close tolerances.

From preliminary tests, stress-strain curves are produced which predict the amount of work-hardening which will result from various amounts of deformation. From these curves the number of pressing operations and intermediate anneals required for the production of a particular component can be assessed, so that tearing or other damage to the metal is avoided.

The main methods of forming sheet-metal components comprise hand beating, spinning, deep-drawing, pressing and rolling. The various techniques are self-descriptive and need little explanation here. The sheet is made to take up the required shape either to given measurements or with the aid of dies and the design is arranged so that the deformation is not highly localised nor excessive for the metal.

Since many of the high-temperature alloys readily work-harden, it is necessary to carry out the shaping of the component with a

greater number of stages and intermediate annealing treatments than the more usual sheet materials. After the final stage a thorough inspection is made for any cracking which may have occurred, especially where the deformation may have been most severe.

In the early gas turbines many of the sheet-metal parts were built up from sections and joined by seam welds. Since little was known about the behaviour of welds under the action of heat and stress, the life of such a construction was usually short, severe warping occurring due to the uneven distribution of stress. With the requirements of increased production, however, and as a result of the more intensive study of the new alloys, such methods of construction have now largely given way to pressing and deep-drawing, thus producing a single-piece article with no welded joints. By proper selection of sheet thickness, and the provision of corrugations, ribs and fins for reinforcement and the relaxation of expansion or contraction stresses, the service life of these components has been greatly increased.

Machining and Grinding

The high cost and short useful life of the dies required for the forging of the high-strength alloys has led to a greater use of machines, frequently capable of automatic operation, for cutting or grinding-out blade forms with high precision. The machine may start with the material either as a rough forging or as bar stock and usually deals with more than one part at the same time. Though the alloys are generally not readily machinable with the ordinary tool steels, the use of carbide-tipped tools and research into the most favourable cutting speeds and depth of cut have made the operation rapid and economic.

After machining it is necessary to polish the surface of the blades to remove the slight unevenness, since a small machining groove may easily initiate a fatigue or creep failure. Consequently several methods of surface treatment are used to ensure a perfectly smooth finish. These include sand blasting, hand polishing and electrolytic polishing, together with various patented processes. Of the latter, the barrelling process (as developed by Rotofinish) is widely used for de-burring and generally smoothing down the surface of precision cast parts. The articles to be treated are packed into cylindrical containers together with a special abrasive mixture and the cylinders are then tilted and rotated for a number of hours. There are two types of abrasive mixture used, one wet and the other dry, according to the degree of polishing required. Both contain abrasive compounds blended with fibrous wood and limestone or granite chippings specially treated to give the optimum polishing

conditions. Articles treated in this way do not undergo any appreciable dimensional changes.

In electrolytic polishing use is made of the fact that when a rough surface is made the anode in an electrolytic cell, the solvent action is slightly greater on the projecting portions. Consequently by careful control of the conditions and the use of suitable electrolytes, this selective action proceeds until the surface is perfectly level. The process is only satisfactory when the initial surface is already relatively smooth but it produces a surface free from distortion, as opposed to mechanical polishing.

Extrusion

The difficulties arising in the production of cooling passages in hot-worked blade material have now been successfully overcome by the manufacturers of the Nimonic alloys and involves the use of a special extrusion process. Prior to extrusion the billet is drilled with a number of suitably distributed holes which are then filled with rods of a material having deformation characteristics similar to the Nimonic alloy at the extrusion temperature. After extrusion to a roughly aerofoil section, the bar is finished by a rolling operation. The filler material is then removed by chemical action, leaving small passages through the blade section.

For creep-resisting alloys, extrusion has an important advantage as a hot working process since the operation takes only a few seconds and consequently all the forming is carried out at the most suitable temperature. The application of extrusion is, however, rather restricted since shaping can obviously only be carried out on the cross-section of the component. There is also a tendency for the degree of hot-working to be variable along the length of the extrude. Extrusion is, however, proving suitable for the production of ring and flange sections for the gas turbine.

Welding

The aim of a welding operation may be defined as the fastening together of two pieces of metal in such a way that the junction will be as nearly homogeneous as possible, and the strength of a good weld should be comparable with that of the joined metals. In the turbine, welded joints are to be found mainly in the construction of rotor assemblies, ducting, combustion chambers and heat-exchangers. Wherever possible the joined metals are of the same composition and the weld metal is also selected to be as similar in analysis and properties as is consistent with the requirements of the welding operation. In certain cases, however, it is very advantageous to weld together different types of steel, particularly ferritic

to austenitic, and although this constitutes a very difficult problem, it has been successfully carried out in the construction of composite rotors.

Welding methods fall into two basic categories, namely "forge and resistance" and "fusion" welding. In the first group are all the methods which consist of bringing the two metals together under pressure while heat is applied in some way so that the metals become sufficiently plastic to join together, e.g. hammer, resistance, spot and butt welding. Fusion welding, on the other hand, concerns the actual melting of the metal in the region of the join, adding further metal to complete the union if required. It is typified by gas, carbon or metallic arc, and argon arc welding.

It is not necessary to discuss here the practical details of these various welding techniques but the effects of a welding operation on the surrounding metal and the performance of welded joints at room or elevated temperatures will be considered.

The majority of the steels used at or just above room temperature are of the Cr-Mo, Cr-Mo-V and Mo-V types. The only difficulty usually encountered with these materials lies in preventing the formation of small heat cracks in or near the weld. During welding, hydrogen produced by decomposition of atmospheric moisture may also enter the molten metal in such quantities that on cooling a large amount of gas evolution occurs at the moment of solidification. This results in an unsound joint and the only method of avoiding this lies in adequate shielding of the welding operation, either by suitable fluxes or a flow of inert gas.

Certain of the alloy steels may tend to air-harden on cooling from the welding temperatures. This will result in stresses being set up in the metal which may lead to the formation of cracks. This can frequently be avoided by pre-heating the metal prior to welding and cooling slowly afterwards.

Probably the main trouble experienced with welded joints in service is the formation of graphite flakes in the zones adjacent to the weld. This process, known as "graphitisation", has been discussed in Chapter Three. It is less likely to occur when the steel contains appreciable amounts of chromium, molybdenum or vanadium, since these elements stabilise the carbide phase and prevent its decomposition to graphite and ferrite.

Since the wide range of stainless and heat-resisting steels frequently enter into welded constructions in a gas turbine, it is of interest to consider how the composition of these materials may influence the welding operation. The steels fall into two groups, ferritic and austenitic, and these two structures differ widely in their properties.

The plain chromium ferritic steels are liable to undergo grain growth when heated to very high temperatures and the formation of a very coarse-grained structure of this type is very undesirable. This tendency is greatly accentuated by prior cold straining of the alloy, such as the forming of sheet materials generally involves. In addition weld deposits in certain composition ranges tend to be of a very brittle nature, so that unless this characteristic is removed by a thorough final heat-treatment such welds are unsuitable for use under highly stressed conditions.

It is for such reasons as these that the welding of ferritic stainless steel is generally avoided wherever possible.

An austenitic alloy, on the other hand, is usually readily weldable by every method except the simple forge welding. This exception arises from the fact that on heating the alloys form protective oxide films which interfere with the joining of the metal surfaces.

The formation of the sigma phase, discussed in Chapter Three, may occur in welded joints during service. It has been found that the occurrence of micro-cracks in austenitic weld deposits can be avoided by ensuring that a proportion of the delta-ferrite phase is formed. This constituent, however, transforms to sigma very readily if the temperature of service exceeds about 550°C (1000°F) and the presence of the sigma constituent may make the material brittle and liable to fracture. Now that the properties of these various phases are better understood it is generally possible to arrange the composition of the weld metal so that, while fulfilling the necessary strength requirements, the formation of sigma in dangerous amounts is avoided.

The large amount of heat generated during fusion welding and the relatively slow dissipation of this heat in thick sections of metal may give rise to a large grain size in the filler metal and adjacent material. This may be removed by re-heating to 800–900°C (1475–1650°F) so that on cooling recrystallisation occurs and normalises the grain size. This heat-treatment also serves to stress-relieve the weld, for which a temperature of 600–650°C (1110–1200°F) is required.

The occurrence of "weld decay", discussed in Chapter Three, is significant only where corrosion resistance is likely to be required.

It is sometimes necessary to join an austenitic to a ferritic alloy. Wherever possible the use of welding for this purpose is avoided, since considerable differences in thermal expansion coefficients are involved and a heat-treatment for one type may be unsuitable for the other. In addition, in service there may be a strong tendency for the carbon of the ferritic steel to diffuse through the weld to the austenitic steel.

However, the joining by welding of these two materials has been successfully accomplished in several gas turbine applications where the advantages greatly outweigh the disadvantages. The use of a built-up rotor consisting of austenitic alloy at the hotter end and ferritic in the cooler regions confers an appreciable saving in cost and it is to be expected that more use will eventually be made of this type of construction. The General Electric Co. of America has made considerable use of a composite disc having an austenitic rim of Timken 16-25-6 alloy welded to a ferritic central portion, generally a nickel-chromium-molybdenum steel. A good many turbines are running with such a disc construction and no troubles appear to arise from the weld.

Failure in such composite welds would generally be expected to arise from the stresses set up by the different thermal characteristics of the two types of alloy. This must be minimised by selecting materials with the closest possible similarity in this respect. The diffusion of carbon across the weld, mentioned above, can be restrained by the addition of strong carbide-forming elements, such as titanium and columbium, to the ferritic steel and by keeping the temperature of the junction in service as low as possible.

The practice of welding axial compressor or turbine blading to the discs is rarely adopted. This would be advantageous in avoiding complicated root fixings and their associated stress concentration problems but would involve sacrificing the considerable damping of vibrations achieved with loose fixings. Since gas turbine alloys have only low inherent damping capacities, the materials themselves cannot adequately damp out the induced fluctuating stresses. A welded joint therefore would greatly increase the risks of a fatigue failure.

Heat-treatment

The heat treatment necessary for the production of the maximum creep-resistant properties makes a high-temperature alloy very difficult to machine and, consequently, has to be left until the required final shape has almost been obtained since a certain amount of surface deterioration is almost inevitable and slight distortion during the treatment may also occur. It is usual, therefore, with small components to heat-treat when the dimensions are a few thousandths of an inch oversize, following the treatment by grinding or electro-polishing to size.

The preliminary solution treatment is generally carried out in the temperature range 1000–1200°C (1830–2200°F) and the time for treatment is usually of several hours' duration depending on the composition of the alloy concerned. The surface oxidation of the

alloy in air at such temperatures would usually be rather severe, so that the use of a protective atmosphere is desirable. The commonest method of producing "neutral" or even slightly reducing atmospheres consists of burning coal-gas with a controlled amount of air. Automatic control equipment is usually fitted, the setting of the burners being adjusted according to signals received from a continuous gas analysis apparatus monitoring the furnace atmosphere; in such cases the furnace must be heated indirectly, generally by electricity.

After the solution treatment the material is cooled, using water, oil or air according to the rate of cooling required. The subsequent precipitation treatment is carried out in similar furnaces at a lower temperature, rapid cooling again generally being necessary.

The essential factor in the mass-production of components in creep-resisting alloys is uniformity of treatment. The development of the maximum high-temperature strength depends on a close control of the processing. The furnace atmosphere, temperature and time of treatment must be kept within the prescribed limits, since appreciable variations will probably produce inferior material. A continuous heat-treatment furnace consisting of a travelling hearth and gradually increasing temperature zones, is often used so that all the components are heated slowly and uniformly to temperature, and maintained there for the correct time.

Pickling and Descaling

In the production of a finished article from a creep-resisting alloy, a large number of mechanical working operations are carried out and frequent annealing treatments are necessary. In each case, surface scale, formed during the time spent in the preheating or annealing furnaces, may tend to become embedded in the surface layers of the alloy during the mechanical working. Consequently it is necessary to clean away the oxidised layers at various stages, as the alloy is reduced from ingot to the finished product.

This may be done by actually machining away the surface layers in cases where the alloy is still in a massive form, e.g. ingots are machined before rolling, to remove surface impurities. In smaller sizes a chemical treatment will be applied which serves to loosen the scale and allow it to be washed or mechanically freed from the alloy. A wide variety of solutions are used for this operation according to the composition of the alloy and the scale, since even on the same alloy the nature and adhesion of the scale may be influenced by the atmosphere used in the furnaces. It is essential that the solution should not dissolve the alloy to any appreciable extent, although attacking rapidly the region of the scale-metal contact.

The most frequently used pickling baths contain dilute solutions of the mineral acids, sulphuric, nitric, hydrochloric and hydrofluoric, either singly or as mixtures, maintained at temperatures of the order of 50–70°C (120–160°F). To these solutions various organic compounds such as pyridine or size may be added where experience shows that they aid in inhibiting metal attack.

Other alloys may be more easily cleaned by the use of a fused salt bath, for which there are several patented compositions. A particularly good descaling process, entailing no metal loss whatever, uses a 2% solution of sodium hydride in molten caustic soda (I.C.I. patent). The hydride is formed by bubbling hydrogen through a special chamber in which molten sodium floats on the surface of the sodium hydroxide. The hydrogen combines with the sodium, the hydride so formed passing into solution in the hydroxide and diffusing into the main chamber. When oxidised material is immersed in the bath, the hydride rapidly reduces the oxides back to metal. Subsequently a cold-water plunge detaches the reduced scale by thermal shock. Stringent safety precautions are required in the operation of this type of bath, since the materials used are strongly corrosive and caustic, and in the case of the sodium and hydrogen, highly inflammable.

These intermittent cleaning operations during manufacture are known as "pickling", while the term "descaling" is generally reserved for the final cleaning in which it is desired to provide a surface entirely free from scale for inspection. For this final treatment the solutions used must produce a very clean surface free from pitting and consequently may dissolve more metal than the pickling solutions.

In the case of aluminium alloys a special treatment known as "anodising" is frequently employed as a final finishing operation. The component is made the anode in an electrolytic cell containing a dilute acid, generally sulphuric or chromic, when a very thin film of aluminium oxide is formed. This film is found to have a beneficial effect on the resistance to fatigue, possibly an indirect result of the greatly improved corrosion resistance.

Inspection and Testing

Before a component is put into service under stress conditions, it is usually subjected to one or more non-destructive tests to detect any faults which cannot be found by visual inspection. The thoroughness of the investigation will clearly depend to some extent on the service requirements and the inspection methods will vary with the dimensions of the component and the type of fault likely to be encountered.

It is now a common practice to supplement the visual inspection, which will only detect large cracks or inclusions at the surface, by x-ray or gamma-ray radiography. In each case the result is recorded on film and the radiograph examined for signs of internal faults. The cost of films and processing can sometimes be avoided by the use of fluorescent screens, but this is generally restricted to the light alloys where the penetration of the radiation is sufficiently easy to render faults visible by such a method.

For the detection of very small cracks either at the surface or immediately underneath a number of techniques are in use, including:

(a) methods utilising the interruption or variation of electrical and magnetic measurements by the presence of faults in the metal.
(b) the absorption and subsequent seepage of special fluids, e.g. coloured or fluorescing compounds, from the crack.
(c) the generation of a strong magnetising field followed by the application of magnetic particles, suspended in paraffin. The particles will adhere to the metal in the region of cracks or inclusions, since strong local magnetic fields are set up at such points. This method is, of course, restricted to magnetic alloys.

The detection of faults in the body of a casting or forging is more difficult, especially in cases where the thickness of metal is too great for radiographic techniques. In recent years, however, the utilisation of ultrasonic waves has given excellent results. The method is based on the fact that if ultrasonic elastic waves are generated in a metal, they will travel forward until reflected by a surface or internal fault. Consequently in sound metal the waves reflected from the opposite surface will be detected after a constant time interval whereas the presence of an internal crack will be indicated by a second reflected wave being received before the main one.

In practice the special electrical properties of a quartz crystal are utilised for the production of the waves, since the application of a rapidly alternating potential results in the sympathetic expansion and contraction of the crystal. If a specially cut quartz crystal is placed in good contact with the metal and an alternating potential, usually with a frequency in the range 20 Kc/s to 500 Mc/s, applied ultrasonic waves will be set up and transmitted through the metal. After reflection from the opposite surface a second quartz crystal can be used to pick up the vibrations and transform them into electrical impulses, these being examined by a cathode ray oscilloscope. Since the waves would also be transmitted across the surface

layers of the metal to the detecting crystal and mask the reflections, it is customary to arrange for the two crystals to operate intermittently so that the receiver only operates to pick up the waves which have traversed the greater path. In more recent instruments the same crystal is made to perform both functions alternately.

By observing the pattern produced on the oscilloscope a trained operator can readily observe the slight variations which indicate the presence of a fault. In addition, an estimate of the location can generally be made.

The application of the technique is capable of further development but already provides a valuable inspection method for both production components and materials which have been in service for some time under conditions liable to cause internal cracks. The necessity for a good surface finish for the successful use of this technique can be overcome to some extent by the use of an oil film between the crystal and the metal. Curved crystals may be used in the case of cylindrical surfaces.

A further inspection method devised by Power Jets Ltd., which has been used for turbine blades, consists of the accurate measurement of the natural frequency of vibration of the new blade and at intervals during service. The value is dependent on the nature of the alloy and the dimensions of the component. Consequently the development of any internal fault will alter the natural frequency. By recording the results obtained with each blade before being put into service and comparing them with subsequent measurements, the development of faults can be detected before they become apparent to visual inspection.

Refractory Materials

The application of refractory materials in the gas turbine is assuming more importance with the development of units burning low-grade fuels. The use of a refractory brick lining in the combustion chamber can be very valuable in permitting a higher flame temperature and encouraging more uniform combustion. In cases where the large mass involved in the use of brickwork is undesirable, a great deal can also now be done to reduce the operating temperatures of metal components by the use of ceramic coatings.

The refractory materials are mainly used in three conditions, as brickwork, plastic refractory mixtures, or coatings, according to the particular requirements.

Refractory Brickwork

The materials used in the manufacture of refractory bricks in general are very numerous but since in the combustion chamber of the

industrial gas turbine it is necessary for the bricks to withstand extremely high temperatures, the choice is restricted to a few materials, e.g. silica, the alumino-silicates, magnesite or chrome-magnesite. The bricks should have the following characteristics:

1. Good thermal shock resistance.
2. A reasonable degree of mechanical strength, particularly when hot.
3. Poor thermal conductivity.
4. A low coefficient of thermal expansion.
5. Structural stability at the maximum service temperature.

To some extent 1 and 4 are incompatible and a suitable compromise has to be made.

In addition, if the products of combustion contain ash particles, resistance to slagging and erosion will be required.

Refractory brickwork has so far been used mainly in conjunction with the low-grade fuels such as the residual oils, coal-tar fuels, peat and coal. Since silicious materials are more prone to slagging in the presence of the ashes of these fuels, the combustion chambers have generally been lined with the alumina-rich refractories, such as sillimanite (60–80% Al_2O_3). This type of brick is made by grinding calcined natural sillimanite to a suitable degree of fineness, bonding this with a little clay and firing after moulding to shape. The resultant brick is extremely stable at temperatures up to 1500–1600°C (2732–2912°F) and has a good resistance to thermal shock. The coefficient of expansion is low, being of the order of 0·5% from room temperature to 1000°C (1832°F).

Plastic Refractory Mixtures

In certain designs of combustion chamber it may be that the standard brick shapes are unsuitable. In such cases the use of the plastic or castable refractories may be advantageous. They consist of finely ground refractories combined with a small amount of a plastic clay. Special brick shapes can thus be made in small quantities, permitting interlocking arrangements, etc., by the user.

A similar type of material is used as a refractory mortar. This is used in refractory brick construction in much the same way as ordinary mortar. As well as binding the bricks together, gas-tight joints are more easily obtainable. When slagging is taking place on the hot face of the bricks, a good mortar can prevent any premature failure of the wall due to the penetration of liquid slag between the bricks. It is clear that the high-temperature properties of the mortar have to be in no way inferior to those of the bricks.

Refractory Coatings

The uses and properties required in a refractory coating have been discussed in Chapter Five. Various mixtures of ceramic materials have been produced for coating such components as the flame tubes or jet pipes of aircraft gas turbines but the same basic method of production applies to them all. The method consists essentially of:

1. Preparation of the ceramic coating material.
2. Treatment of the component prior to coating.
3. Application of the coating and firing.

The coating generally consists of metallic oxides and fluxes in powder form. These are thoroughly mixed and smelted. When molten the liquid is allowed to drop into cold water when a glassy product, known as a "frit", is formed. This is then dried and reground. During the grinding a small amount of a binding agent is added, to assist in the coating operation.

The component to be coated must be thoroughly cleaned and the surface slightly roughened to increase the adhesion of the coating. The frit is then applied either by dipping or spraying and the coating first dried carefully and then fired.

APPENDIX OF COMPOSITIONS

The following table gives the approximate composition of the various alloys mentioned in this book. It is not intended to be a complete list of heat-resisting materials and it should be noted that a particular type of alloy may be produced under several different trade names

BRITISH ALLOYS

TYPE	EXAMPLES	COMPOSITION
		Non-ferrous Materials
Aluminium Alloy	Hiduminium R.R.57	6% Cu, 0·25% Mn, 0·1% Ti
	Hiduminium R.R.58	2·2% Cu, 1·5% Mg, 1·0% Fe, 1·2% Ni, 0·1% Ti
	Hiduminium R.R.59	2·2% Cu, 1·5% Mg, 0·85% Si, 1·0% Fe, 1·2% Ni, 0·1% Ti
Aluminium Bronze	Langley Alloys Hidurax 1/12A	As per D.T.D.197A. 8–11% Al, 4–6% Fe, 4–6% Ni
Titanium Alloy	Titanium 371	Ti-Sn-Al alloy
Magnesium Alloy	Elektron Z.5.Z	0·7% Zr, 4·5% Zn
	Elektron R.Z.5	0·7% Zr, 4·0% Zn, 1·3% Rare Earths
	Elektron Z.R.E.1	0·6% Zr, 2·2% Zn, 2·7% Rare Earths
	Elektron M.C.Z.	0·6% Zr, 3·0% Rare Earths
	Elektron Z.T.1	0·7% Zr, 2·2% Zn, 3·0% Thorium
Copper-nickel Alloys	Monel	67% Ni, 30% Cu, 1·4% Fe, 1·0% Mn
	K-Monel	66% Ni, 29% Cu, 2·75% Al, 0·9% Fe, 0·4% Mn, 0·25% Si, 0·15% C

BRITISH ALLOYS (*continued*)

TYPE	EXAMPLES	COMPOSITION
		Ferrous Materials
Mo-V	Mova	0·2% C, 0·25% Si, 0·5% Mn, 0·7% Mo, 0·25% V
Cr-Mo-V	H.51	0·2% C, 0·3% Ni, 0·75% Cr, 0·5% Mo, 0·25% V
1% Cr-Mo	H.31	0·4% C, 0·3% Si, 0·5% Mn, 1·0% Cr, 0·75% Mo
1% Cr-Mo-V	Hecla 153B	0·4% C, 0·3% Si, 0·5% Mn, 1·0% Cr, 0·75% Mo, 0·25% V
3% Cr-Mo		0·2% C, 0·2% Si, 0·7% Mn, 0·4% Ni, 3·0% Cr, 0·5% Mo
3% Cr-Mo-V		0·4% C, 0·3% Si, 0·6% Mn, 0·4% Ni, 3·0% Cr, 1·0% Mo, 0·25% V
3% Cr-Mo-V-W	Chrom Va-W H.40	0·2% C, 0·3% Si, 0·4% Mn, 3·0% Cr, 0·5% Mo, 0·75% V, 0·5% W
Silchrome	H.3.A	0·57% C, 1·2% Si, 6·0% Cr, 0·5% Mo
12% Cr	F.G.	0·2% C, 0·4% Si, 0·3% Mn, 13% Cr
12% Cr	Hecla HGT.4	0·17% C, 0·5% Si, 1·0% Mn, 11·5% Cr, 0·6% Mo, 0·2% Cb, 0·2% V, 0·07% N_2, 0·3% B
12% Cr	H.46	0·2% C, 0·2% Si, 11·5% Cr
12% Cr	Silver Fox 11	0·2% C, 12% Cr
12% Cr	G.2	0·5% C, 1·3% Si, 0·7% Mn, 13% Ni, 13% Cr, 0·1% Mo, 2·5% W
18% Cr, 10% Ni	R.16	0·1% C, 1·2% Si, 0·8% Mn, 10% Ni, 18% Cr, 0·7% Ti
	F.C.B.(T)	0·1% C, 0·6% Si, 1·5% Mn, 12% Ni, 18% Cr, 1·2% Cb
High-Alloy Steels	R.20	0·15% C, 0·3% Si, 0·8% Mn, 14% Ni, 19% Cr, 1·7% Cb
	Stayblade	0·2% C, 1·0% Si, 0·6% Mn, 8% Ni, 20% Cr, 1·2% Ti
	Rex 78	0·1% C, 0·7% Si, 0·8% Mn, 18% Ni, 14% Cr, 4·0% W, 0·6% Ti, 4·0% Cu
	G.40	25% Ni, 20% Cr
	G.56	23% Ni, 13% Cr
	Red Fox 31	15% Ni, 25% Cr
	Red Fox 33	30% Ni, 20% Cr

H.R. Crown Max.	0·2% C, 1·5% Si, 0·4% Mn, 12% Ni, 23% Cr, 3·0% W
Immaculate 5	0·12% C, 1·5% Si, 0·8% Mn, 21% Ni, 25% Cr
Rex 326	0·25% C, 0·8% Si, 3·0% Mn, 17% Ni, 16% Cr, 7·0% Co, 3·0% Mo, 2·0% Cb
G.18B	0·4% C, 1·0% Si, 0·8% Mn, 13% Ni, 13% Cr, 10% Co, 2·0% Mo, 2·5% W, 3·0% Cb
G.19	0·4% C, 1·0% Si, 0·8% Mn, 13% Ni, 19% Cr, 10% Co, 1·8% Mo, 2·5% W, 3·0% Cb
G.42B.	15% Ni, 19% Cr, 25% Co, 32% Fe
Hecla E.M.35	0·35% C, 0·4% Si, 1·5% Mn, 15% Ni, 17% Cr, 12% Co, 2·9% Mo, 2·5% W, 1·0% Cb, 0·09% N_2
Hecla M.M.35	0·35% C, 0·5% Si, 1·5% Mn, 20% Ni, 21% Cr, 20% Co, 2·9% Mo, 2·5% W, 1·0% Cb, 0·12% N_2
Hecla E.M.20	0·18% C, 0·4% Si, 1·5% Mn, 15% Ni, 17% Cr, 12% Co, 2·9% Mo, 2·5% W, 1·0% Cb, 0·09% N_2
Hecla M.M.20	0·18% C, 0·5% Si, 1·5% Mn, 20% Ni, 21% Cr, 20% Co, 2·9% Mo, 2·5% W, 1·0% Cb, 0·12% N_2
	Cobalt Base Alloys
G.32 G.34	0·27% C, 0·5% Si, 0·8% Mn, 10·5% Ni, 19% Cr, 46·6% Co, 2·2% Mo, 1·4% Cb, 3·0% V, bal. Fe
	Nickel Base Alloys
Nimonic 75	0·08%–0·15% C, 0·2–0·6% Ti, 18–21% Cr, 1% max. Si, 1% max. Mn, 5% max. Fe, 0·5% Cu, bal. Ni
Nimonic 80	0·1 % max. C, 1·8–2·7% Ti, 18–21% Cr, 0·5–1·8% Al, 1% max. Si, 1% max. Mn, 5% max. Fe, 2% max. Co, bal. Ni
Nimonic 80A	0·1% max. C, 1·8–2·7% Ti, 18–21% Cr, 0·5–1·8% Al, 1% max. Si, 1% max. Mn, 5% max. Fe, 2% max. Co, bal. Ni

BRITISH ALLOYS (*continued*)

TYPE	EXAMPLES	COMPOSITION
		Nickel Base Alloys (continued)
High Alloy Steels	Nimonic 90	0·1% max. C, 1·8–3·0% Ti, 18–21% Cr, 0·8–2·0% Al, 1·5% Si, 1% max. Mn, 5% max. Fe, 15–21% Co, bal. Ni
	Nimonic 95	0·15% max. C, 2·3–3·5% Ti, 18–21% Cr, 1·4–2·5% Al, 1% max. Si, 1% Mn, 5% Fe, 15–21% Co, 0·5% Cu, bal. Ni
	Nimonic 100	0·3% C, 1–2% Ti, 10–12% Cr, 4–6% Al, 4·5–5·5% Mo, 0·5% Si, 2% Fe, 18–22 Co, bal. Ni
	Nimonic C	as Nimonic 80
	Nimonic D	2% C, 37% Ni, 18% Cr, bal. Fe
	Nimonic F	as Nimonic 75 but higher Fe content

MANUFACTURERS OF THE BRITISH ALLOYS

T. Firth & J. Brown Ltd.	Mova, Chrom Va-W
Firth Vickers Stainless Steels Ltd.	F.G., F.C.B.(T), Stayblade, Rex 78, H.R. Crown Max., Immaculate 5, Rex 326
Jessop-Saville	H.51, H.31, H.40, H.3.A, H.46, G.2, R.16, R.20, G.40, G.56, G.18B, G.19, G.42B, G.32, G.34
Hadfields Ltd.	Hecla 153B, H.G.T.4, Hecla E.M.35, Hecla M.M.35, Hecla E.M.20, Hecla M.M.20
Samuel Fox & Co., Ltd.	Silver Fox 11, Red Fox 31 and Red Fox 33
Henry Wiggin & Co., Ltd.	All Nimonic alloys, Monel and K-Monel

MANUFACTURER	EXAMPLES	COMPOSITION
		Ferrous Materials
Union Carbide	N-155 (low C)	0·2% max. C, 1·0% max. Si, 1–2% Mn, 21% Cr, 20% Ni, 2·75–3·75% Mo, 20% Ti, 2–3% W, 1% Cb
Universal Cyclops	19-9 DL	0·3% C, 0·6% Si, 1·1% Mn, 19% Cr, 9·0% Ni, 1·25% Mo, 0·3% Ti, 1·2% W, 0·4% Cb
Timken	16-25-6	0·1% C, 1·35% Si, 0·7% Mn, 16% Cr, 25% Ni, 6% Mo, 0·15% N_2
Westinghouse	Discaloy 24	0·03% C, 0·64% Si, 0·92% Mn, 13·1% Cr, 25·5% Ni, 2·83% Mo, 1·83% Ti, 0·16% Al
Alleghany-Ludlum	S.590	0·4% C, 0·7% Si, 1·5% Mn, 20% Cr, 20% Ni, 4% Mo, 20% Co, 4% W, 4% Cb
Westinghouse	K.42B	0·05% C, 0·7% Si, 0·7% Mn, 18% Cr, 43% Ni, 2·5% Ti, 22% Co, 0·2% Al, 13% Fe
		Cobalt Base Alloys
Alleghany-Ludlum	S.816	0·38% C, 0·7% Si, 1·5% Mn, 20% Cr, 20% Ni, 43% Co, 4·0% Mo, 4·0% W, 4·0% Cb, 3% Fe
Haynes-Stellite	Vitallium Haynes-Stellite 21	0·25% C, 0·6% Si, 0·6% Mn, 27% Cr, 3% Ni, 62% Co, 5% Mo, 1% Fe
	Haynes-Stellite 23	0·4% C, 0·6% Si, 0·3% Mn, 24% Cr, 2·0% Ni, 66% Co, 6·0% W, 1·0% Fe
	Haynes-Stellite 27 6059	0·4% C, 0·6% Si, 0·3% Mn, 25% Cr, 32% Ni, 34% Co, 6·0% Mo, 1·0% Fe
	Haynes-Stellite 30 422-19	0·4% C, 0·6% Si, 0·6% Mn, 24% Cr, 16% Ni, 51% Co, 6·0% Mo, 1·0% Fe
	Haynes-Stellite 31 X-40	0·4% C, 0·6% Si, 0·6% Mn, 25% Cr, 10% Ni, 55% Co, 8·0% W, 1·0% Fe

AMERICAN ALLOYS (*continued*)

MANUFACTURER	EXAMPLES	COMPOSITION
		Nickel Base Alloys
International Nickel Co.	Inconel	0·04% C, 0·2% Si, 0·35% Mn, 14–15·5% Cr, 76% Ni, 6–7% Fe
	Inconel X	0·04% C, 0·3% Si, 0·7% Mn, 15% Cr, 73% Ni, 0·9% Cb, 7·0% Fe, 2·5% Ti, 0·9% Al
	Kinsalloy B	70% Ni, 22% Mo, 8·0% Al
Haynes-Stellite	Hastelloy A	0·1% C, 0·7% Si, 0·2% Mn, 59% Ni, 20% Mo, 20% Fe
	Hastelloy B	0·1% C, 0·7% Si, 0·8% Mn, 1·0% Cr, 65% Ni, 28% Mo, 5% Fe
	Hastelloy C	0·1% C, 0·7% Si, 0·8% Mn, 16% Cr, 57% Ni, 17% Mo, 4·0% W, 5·0% Fe
	Hastelloy D	0·1% C, 10% Si, 1·0% Mn, 85% Ni, 1·0% Fe, 4·0% Cu

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